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# COLUMBIUM AS A MICRO-ALLOYING ELEMENT IN STEELS AND ITS EFFECT ON WELDING TECHNOLOGY

SSC-154

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BY  
T. M. NOREN

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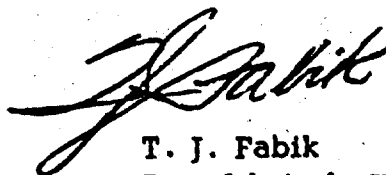
30 August 1963

Dear Sir:

Dr. T. M. Norén, of Oxelösunds Järnverk, Oxelösund, Sweden, accepted the invitation to participate in the Annual Meeting (held on March 14 and 15, 1962 in Washington, D. C.) of the Committee on Ship Steel of the National Academy of Sciences-National Research Council, one of the principal advisory committees to the Ship Structure Committee. The enclosed report entitled Columbium as a Micro-Alloying Element in Steels and its Effect on Welding Technology was prepared by Dr. Norén to summarize his remarks for the Committee on Ship Steel.

Please send any comments on this report addressed to the Secretary, Ship Structure Committee.

Yours sincerely,



T. J. Fabik  
Rear Admiral, U. S. Coast Guard  
Chairman, Ship Structure  
Committee

SSC-154

Special Report

on

COLUMBIUM AS A MICRO-ALLOYING ELEMENT IN STEELS  
AND ITS EFFECT ON WELDING TECHNOLOGY

by

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Washington, D. C.  
U. S. Department of Commerce, Office of Technical Services  
August 30, 1963

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## WELDABILITY AS A METALLURGICAL CONCEPT - A DEFINITION

As a result of the heat influence to which a steel is exposed in any form of welding, the material undergoes certain changes, some of which are permanent. These changes may occur as microstructure transformations during the cycle of heating and cooling, or as changes in shape or dimensions due to thermal stresses. A steel which can be welded without application of complicated precautions to avoid dangerous consequences of these changes regarding the stability of the welded structure is said to possess good weldability.<sup>1</sup>

If, for certain steels, on the other hand, a normal welding process will imply serious danger of causing failure in a welded component due to the changes mentioned or if actual defects, such as cracking, occur already during welding or immediately after, certain precautions must be taken or special pre- and/or post-treatments carried out. Such steels are said to possess limited weldability.<sup>1</sup>

The term "unweldable steels" is not realistic. Any steel can be welded provided correct metallurgical conditions are chosen. Sometimes, however, these conditions may be impossible to realize in practical production work. The rapid heating and cooling cycles applied to a steel by welding may be characterized as a thermal shock influence or a series of such influences on the steel. The weldability grade can be regarded as the ability of the steel to withstand this thermal shock attack.

The weldability concept is complex and therefore difficult to define. Still it is one of the commonly used metallurgical terms. It is indeed not quite clear what is meant, in daily talk, by a weldable steel. Moreover the word "weldability" has a limited range of meaning and refers only to the base metal itself and how this will react during a welding process. Consequently, there is a need for another concept including the whole welded joint and how its properties will influence the stability of the welded structure. Therefore, a term like "function stability" of welded joints is more adequate and includes the weldability of the base metal as an important and necessary but not complete determination of the expression.<sup>2</sup> There is a weld metal, too, in the welded joint, the properties of which are more or less dependent on the composition of the base material.

## WELDABILITY AS A PROBLEM COMPLEX IN STEEL METALLURGY

The weldability concept in the widest sense can be almost completely covered by some main groups of metallurgical phenomena which can be said to have an influence on the ability of a steel to undergo necessary welding technology treatment. These groups refer to melting and solidification as well as to microstructure transformations, temperature-dependent mechanical properties, corrosion and oxidation phenomena or, in other words, to several possible changes in physical and chemical behavior. Detrimental changes which can be expected to occur under certain circumstances may be divided into seven groups and can be summarized as follows:

1. Longitudinal weld cracking (solidification cracks or "hot cracks").
2. Transversal weld cracking (shrinkage cracks or "cooling cracks").
3. Hardening embrittlement in the weld or the transformation zone of the steel.
4. Normal brittle behavior of the weld or the steel below a characteristic critical temperature and under severe stress conditions, e.g. residual welding stresses.
5. Embrittlement due to microstructure instability of the weld or the steel at low and medium temperatures.
6. Embrittlement due to microstructure instability of the weld or the steel at high temperatures.
7. Decrease of corrosion and oxidation resistance of the weld or the steel due to residual welding stresses and/or certain microstructure formations.

Various metallurgical phenomena may be the cause of a defect or change in properties, which is characteristic for each of the seven groups, but within a certain group these phenomena are rather closely connected as such. Hence a complete weldability investigation of a steel should be performed with regard to these seven groups and a development of a new steel type for welded structures should accordingly be directed towards properties of the steel which will not contribute to what is mentioned under the seven groups within the anticipated fabrication and service conditions.

This can easily be stated and is easily

understood but not so easily carried out. The difficulties will appear already before the steel has come into the ladle.

More or less covering all the seven groups mentioned above are the mechanical properties of the steel and how they will change under the influence of possible defects due to welding.

Provided there are no defects forming sharp notches, the yield strength of a steel will increase with falling temperature and the plasticity will decrease accordingly. Independently of the testing method used, one will find that, at certain higher temperatures, a steel will behave in a ductile way and at certain lower temperatures in a mainly brittle way. Between these two temperature ranges there is a transition range within which the fracturing conditions of the material may be a little more complicated.

The more severe the stress conditions are with respect to triaxiality the higher is the temperature at which the transition range between the ductile and brittle behavior of a steel is to be found.

It will be stated (p. 14) that it would not be realistic to expect a welded structure to be free from defects in the welded joints. Such defects will act as most dangerous notches, and on considering the function stability of a welded structure, it is indeed important to bear this in mind. On the other hand, defects in the steel itself which are localized far away from welds may be regarded as being of less importance. The function stability of a welded structure will be determined by the amount and location of defects in and around the welded joints, i.e. in the parts of the structure which are under influence of welding stresses. There are many types of defects in welds or adjacent to a weld, which may be regarded as possible initiation points for a brittle fracture, a few examples of which are shown in Fig. 1-5.

There are a great many various methods by which the tendency of a brittle behavior of a steel at certain temperatures can be determined. The most simple one is the impact testing, which is rather useful provided the strain rate on initiating the fracture at the notch root of the test bar is mainly corresponding with practical circumstances. Such a testing method is the Charpy V-notch testing, too well-known to be described here. However, I would like to quote the conclusion of an investigation

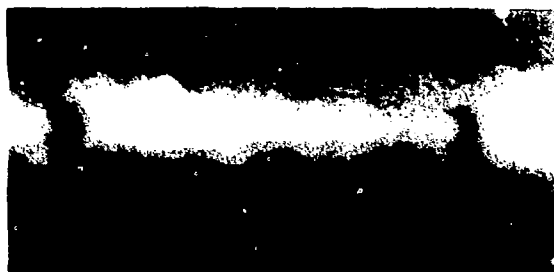


FIG. 1. X-RAY PHOTOGRAPH OF A WELD SHOWING PIPE FORMATION, WHERE CHANGE OF ELECTRODE HAS TAKEN PLACE DURING VERTICAL MANUAL WELDING. IN CROSS-SECTION SUCH PIPES OR SHRINKAGE CAVITIES ARE TO BE REGARDED AS MORE OR LESS PRONOUNCED HOT CRACKS.



FIG. 2. REPAIR WELDING WITHOUT PREHEATING OF A SURFACE DEFECT ON A HALF INCH HIGH TENSILE STRUCTURAL STEEL PLATE. AFTER THE WELDING THE REPAIRED SPOT HAS BEEN GROUND. A HARDENING CRACK ADJACENT TO THE WELD HAS APPEARED IN THE MARTENSITIC TRANSITION ZONE.

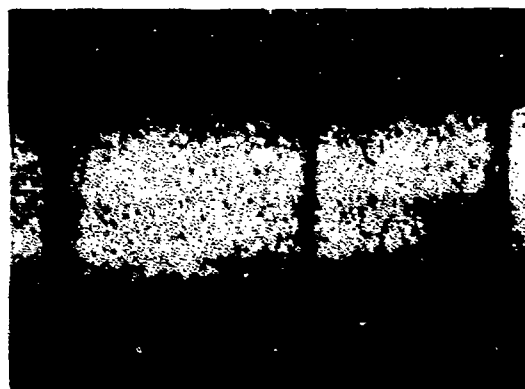


FIG. 3. X-RAY PHOTOGRAPH OF A LOW-ALLOY WELD WITH TRANSVERSAL CRACKS CAUSED BY HYDROGEN EMBRITTLEMENT OF THE PARTIALLY MARTENSITIC WELD METAL. THIS TYPE OF WELD DEFECTS IS ONE OF THE MOST DANGEROUS REASONS FOR INITIATION OF BRITTLE FRACTURES IN THE BASE METAL. EVEN A SLIGHT PREHEATING COULD HAVE PREVENTED THE FORMATION OF SUCH TRANSVERSAL SHRINKAGE CRACKS.

performed by a research committee of Jernkontoret in Stockholm, which states the following about what can be gained by Charpy V-notch testing<sup>3</sup>:

"Below temperatures, corresponding to the lower change of a Charpy V-notch curve, a steel may be expected to behave in a brittle manner under conditions permitting an initiation of a fracture at sufficiently high strain rate.

"At temperatures below the range mentioned residual stresses, e.g. welding stresses, may cause initiation and propagation of brittle fractures, provided sharp notches, e.g. weld defects, are present.

"Above temperatures, corresponding to the lower change of a Charpy V-notch curve, a steel will generally also in practice behave in a ductile and crack-arresting manner."

This corresponds rather well with a British investigation of much the same type<sup>4</sup> and the opinion of G. M. Boyd.<sup>3</sup>

In 1961 Dr. Georg Vedeler<sup>5</sup> presented an excellent report to the Committee on Ship Steel. In this report he states that from a practical point of view the problem of brittle fractures in ships has been solved by the present regulations. He also pointed out that the main problems for the shipbuilders would today be fatigue cracks in the ships.

Concerning fatigue cracking he is no doubt right, but I cannot quite agree with his statement regarding the practical solution of the brittle fracture problem. Doubtlessly he is right by saying that the new regulations have increased safety against brittle service failures. I must admit, however, that so far as I understand there will probably never appear any fatigue crack in a ship that will propagate to an extent that the ship will fail by a fatigue fracture in the conventional meaning. The importance of fatigue cracking in ships or other welded structures is from my point of view that they may act as extremely dangerous initiation points for brittle fractures by their sharp notch effect in parts of the ship, where severe stress conditions occur.

In other words, if the brittle fracture problem is solved from a practical point of view I do not see that fatigue cracking could be of such a great importance. If they are not any longer



FIG. 4. UNDERBEAD CRACKING IN A MARTENSITIC TRANSITION ZONE CLOSE TO THE FUSION LINE OF A WELD IN A LOW-ALLOY STEEL. 1500X



FIG. 5. STRESS CORROSION CRACKS ADJACENT TO A WELD IN AN UNALLOYED STEEL. THE CRACK PATTERN HAS APPEARED IN THE PARTS OF THE BASE METAL WHICH HAVE BEEN UNDER INFLUENCE OF RESIDUAL WELDING STRESSES.

dangerous initiation points for brittle failures, I do not think that in a lifetime of a ship a fatigue crack will get time to extend to such a degree that a residual shear fracture will occur due to overloading of the remaining uncracked area.

On the other hand, if brittle fracture is still a reality, we certainly have to concentrate on fatigue research in connection with welded structures. Fatigue cracking in or around welded joints is probably one of our most dangerous defects to be considered in connection with the function stability of a structure. I would summarize my viewpoints by the following:

1. Defects in welded joints mostly occur in the weld metal itself. Cracking in the transition zone can be more easily overcome.
2. The weld metals of today have normally a very low transition temperature range with regard to brittle fracture. Consequently, the risk for initiation of a brittle fracture in such weld defects is rather limited.
3. Weld defects, however, can easily become the starting point for a fatigue failure, since the fatigue strength under the influence of the notch effect of the weld defect will be very much decreased. If a fatigue crack, starting from a weld defect, extends in a direction where it will reach the surrounding base metal, there is obviously a great risk for initiation of a brittle fracture in the steel, the transition temperature of which may be far higher than that of the weld metal. This might particularly be true with regard to the parts of the base metal under influence of welding stresses.

Vedeler also says in his report<sup>2</sup> that he is inclined to think that for steel with a high-yield point one should have a larger margin to the transition temperature, and the definition of the transition temperature by means of a Charpy V-notch test should be at a higher energy than for ordinary ship steel. Nobody could be more willing to underline this than I am. Some of my own investigations<sup>3</sup> have shown that there is a good reason for stating this.

On the other hand, in case of application high-strength steels to welded structures, and in case we are able to define a transition temperature by, for instance, impact testing that has good relation to practical service conditions, I do not see why we have to fear the high elastic stresses.

Such stresses will no doubt form around welded joints in such steels as Vedeler rightly points out. It is also evident that residual welding stresses must be higher, the higher the

yield strength of the base metal. There is no reason to believe, however, that welding stresses will have another type of influence in a high-strength steel than in an ordinary one.

Many investigators have already shown that above the transition temperature, as we today normally define it, welding stresses will not contribute to brittle failures. In our definition of the transition temperature, as measured by means of the Charpy V-notch test bar, we have already included a certain margin by stating 15-20 ft lbs as a critical impact level for an ordinary mild steel. Therefore when we have found the critical levels corresponding to higher yield strengths and have included a corresponding margin of safety, I definitely believe in the successful application of these new steel types in the welding technology.

For pressure vessels this is already a fact. In connection with shipbuilding I personally think that the main problem is that the modulus of elasticity will still be the same also for the high-strength materials. In principle the brittle fracture problem will become the same whatever the strength of the steel may be.

I would like to finish this part of my report by stating that out from my experience a serious service failure of a welded structure will always culminate in some sort of a brittle fracture, no matter what the foregoing reason may have been--a weld defect, a transition zone crack, a fatigue crack, etc. Therefore, I am not willing to underestimate the importance of studying the brittle behavior of steels for welded structures, in particular the strength of the steel under the influence of the sharpest possible notch (i.e. a natural crack) at low temperatures and under severe welding stress conditions. There are many methods today which can be applied to such studies, one of them being the NC-testing.<sup>4</sup> This method was developed in 1951 for the determination of "the nominal cleavage strength" of a steel surrounding a welded joint (Appendix B). Recent investigations by Pellini and co-workers seem to have followed much the same lines as to the basic ideas about the fracture behavior of steels in relation to the influence of stresses, sharp notches and varying temperatures.

#### MICRO-ALLOY STEELS

From the weldability point of view there is a gap between plain carbon steels and C-Mn-



steels on one side and low-carbon low-alloy steels on the other. We have to find some connection between these two steel groups, and the micro-alloy steels might be what we are looking for. I have already used the term "micro-alloying element" in the title of this paper, and therefore I think I should go straight to the definition of this expression.

What is a micro-alloy steel? It is a steel, the basic composition of which is simply an unalloyed structural steel or, in many cases, a manganese alloy one or even a low-alloy one, to which a small amount of an alloying element has been added--this element having a very strong and sometimes remarkable effect on one or several of the steel properties. On the whole, however, the steel is still characterized by its basic composition as to its general behavior. The amount of micro-alloying elements to be added is one or two powers of ten less than would have been the case for an alloying element in the conventional meaning.

Micro-alloy steels have been used for quite a time. I am thinking of the aluminum-treated steels, in which aluminum certainly does act as a micro-alloying element. Other examples of such elements are vanadium, boron and titanium. Still another is columbium.

As a consequence of the definition of what I have called micro-alloy steels, one can speak about micro-alloy carbon steels, micro-alloy manganese steels, micro-alloy molybdenum steels, etc. The micro-alloying element added to a base composition of a normal type will, as said above, in certain respects change the steel properties more or less drastically but still the steel behavior is mainly dependent on its basic composition.

The influence of micro-alloying elements is by no means the same in all cases. One type of element may change the grain size, another may change the resistance to tempering embrittlement, still another may have an influence on certain transformation reactions during cooling, etc.

The choice of micro-alloying element depends on the property or the properties of the base steel to be stabilized or changed.

Let us now turn to the discussion of columbium as a steelmaking variable, but, before that:

You may regard the statements regarding "weldability" given as a background of this paper as "Elementary, my dear Watson." If so, I quite agree, but then I would only like to make another statement: The simpler you can build the platform on which your research work is based, the better it is. Further, the more systematically you can treat your problems, the safer you feel. Simplicity, senses and systematization must never exclude the necessary brilliance of a successful research work, but will offer you a reasonable safety on applying your results to practice. "Elementary, my dear Watson" - it is all right and I do not care.

#### GENERAL INFLUENCE OF COLUMBIUM AS A MICRO-ALLOYING ELEMENT IN STEEL

Until now there has not been very much written about columbium as a steelmaking variable. Technical information to be found in literature at the moment concerning the behavior of columbium-alloy steels, and information gained by personal contacts with colleagues who have been investigating such steels is limited and contradictory. This seems quite natural since there are probably only a few steel works having had columbium steels in full-scale production. It is our experience that rather few important observations can be made without production experience as to the real influence of columbium as a steelmaking variable.

However, as a basis for the development of columbium-alloy steels and the interest in these steel types, some well-known influences on the properties of steels by the addition of small amounts of columbium in the order of magnitude of 0.01-0.05% have been observed. For instance, there is no doubt that columbium will increase the yield strength of the steel and cause a fine-grained micro-structure.

The increasing yield strength could of course preliminarily be explained by the fine-grained structure, but this does not seem to be the whole truth. An additional effect on the yield strength from columbium itself is probably to be found, but so far as we know, the true reason for this part of the yield strength increase is not definitely explained. It would not be unreasonable to believe, as Ketzer does, that fine-dispersed carbides or perhaps nitrides would in some way or another strengthen the translation planes. Our own investigations are still incomplete and have no contribution

to the solution of this problem to offer.<sup>9</sup> Some observations in the electron microscope may, however, partly confirm the statements of Beiser (Fig. 6-7).

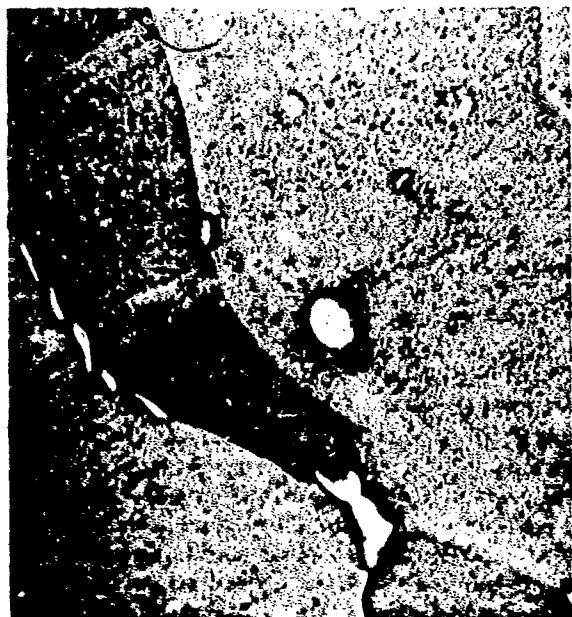


FIG. 6. GRAIN BOUNDARY CEMENTITE AND COLUMBIUM CARBIDE PRECIPITATION IN THE FERRITE OF A CARBON-MANGANESE STEEL WITH 0.02% COLUMBIUM. ELECTRON MICROGRAPH 14,000x.

Even if columbium would have been regarded only as an alloying element which can cause a fine-grained structure in the steel, there would still have been some advantages left for such an addition.

As an oxide former columbium is definitely less strong than for instance aluminum or titanium; alloying elements which are also used for fine-grain treatments. This will permit columbium to be used in fine-grain practice more or less independently of the deoxidation practice applied to the steelmaking. In other words, semikilled (balanced) columbium-treated steels can very well be produced and are in most respects not significantly different from fully killed steels with the same columbium addition, which, of course, has an economic importance.

<sup>9</sup> See, however, Appendix C - a summary of recent investigations by co-workers of the Author.



FIG. 7. GRAIN BOUNDARY CEMENTITE AND COLUMBIUM CARBIDE PRECIPITATION IN THE FERRITE AS WELL AS IN SUBGRAIN BOUNDARIES IN A CARBON-MANGANESE STEEL AS IN FIG. 6 BUT WITH 0.10% COLUMBIUM. THE COLUMBIUM CARBIDE PRECIPITATE IS FAR COARSER THAN IN THE FOREGOING FIGURE. ELECTRON MICROGRAPH 14,000x.

An American patent specification claiming a semikilled columbium-alloy steel appeared as late as November 28, 1961.<sup>10</sup> Some other patent specifications have previously been published in U.S.A. and other countries. Already about 25 years ago columbium was mentioned as an alloying element used in such small additions which are characteristic for today's columbium steels. It might be that even prior to that the effect of columbium was studied. But there seems to have been a pause in the development of these new steel types during a period of about ten years until a new approach to the problem was made simultaneously in various parts of the world.

The need for high-strength steels for welded structures was doubtlessly a main reason. It is well known that many steel works based their interest in the effect of micro-alloying

elements on the fact that for many applications within the welding technology the low-alloy steels were not only involving too great a step towards increased yield strength, but certainly also too high a cost in relation to the steel weight that could possibly be saved without a

drastic change in existing regulations for welded structures. Further, a good deal of the low-alloy steels will not stand the rather rough treatments which can hardly be avoided in most of the welding shops.

With the exception of aluminum, and to a certain extent also boron that is used preferably in combination with low-alloy steels, e.g. molybdenum steels, the use of micro-alloying additions is quite a new field of steel metallurgy. In the invitation letter from Professor Chipman, he asked me to present my most recent thoughts about columbium as a steelmaking variable.

May I say that I have experienced this sometimes confusing alloying element in a way that any correct or, at least, reasonable thought about columbium as a steelmaking variable is indeed recent.

There is very much to be expected in the future concerning our knowledge of this subject, and for the present we have only touched the problem complex which is promising so much. But the solution is still hiding behind a mountain of necessary investigations.

#### METALLURGICAL VARIABLES

In metallurgy we have three variables to apply in order to produce a steel for a given purpose. They are:

1. Composition (including deoxidation practice).
2. Mechanical treatment.
3. Heat treatment.

Our experience of the processing metallurgy of columbium steels is limited to two types of steelmaking processes: the open hearth process and the kaldo process.<sup>1,2</sup> In our full-scale investigations we have studied various types of deoxidation practice, i.e. semikilled steels (balanced steels) as well as silicon-killed steels, and silicon-treated steels with an additional deoxidation by means of aluminum. As examples of what may be called normal columbium steels, I would like to choose the following composition ranges:

C	0.10 - 0.25
Si	0.03 - 0.30
Mn	0.40 - 1.60
Nb	0.005 - 0.05

Within the above-mentioned ranges we have paid most attention to the following three steels, which mainly differ from each other with regard to the carbon contents:

TABLE I  
AVERAGE COMPOSITIONS

	A	B	C
C	0.11	0.22	0.19
Si	0.03	0.03	0.20
Mn	1.0	1.3	1.3
Nb	0.02	0.03	0.03

Steel A has an upper yield strength of about 37 kg/mm<sup>2</sup> (22,625 psi) while steels B and C show a yield point at room temperature of about 47 kg/mm<sup>2</sup> (36,850 psi) as rolled. It should be observed that the difference in composition between steel B and steel C is limited to the carbon contents and the silicon contents. There is an influence of silicon on the strength of the steel, and therefore steel C may be given a slightly lower carbon percentage. The average yield strength increase on adding columbium is 10-12 kg/mm<sup>2</sup> (14,000-17,000 psi).

There is no marked difference between the possibilities of producing about the same quality of steel by the open-hearth process and the kaldo process. However, a kaldo steel, because of its very low oxygen and nitrogen contents, has a superior formability, which will be disclosed by the bending properties, the elongation and the impact properties. This is, however, typical also for unalloyed kaldo steels in relation to the same type of steel made by the open-hearth process and has obviously nothing to do with the columbium addition.

There are several ways of adding columbium to the molten steel. The columbium yield will vary depending upon the adding method used and ranges 50-95%. It is our experience that the highest recovery of a columbium addition will occur on adding the alloying metal (for instance as ferrocolumbium) to the mold.

Portions of ferrocolumbium, each being, for

instance, 10% of the total addition, may be thrown into the mold and will give a yield of 60-70%. Another way of adding ferrocolumbium is an injection method, which we have developed and which we have found will give the highest yield, 90-95%. According to this method a rather fine-grained powder (average grain size about 2 mm) is blown into the steel stream by means of equipment shown in Fig. 8. This method is of course not only connected with columbium additions but with any addition of micro-alloying elements, which on the whole can be added at this stage of the process.

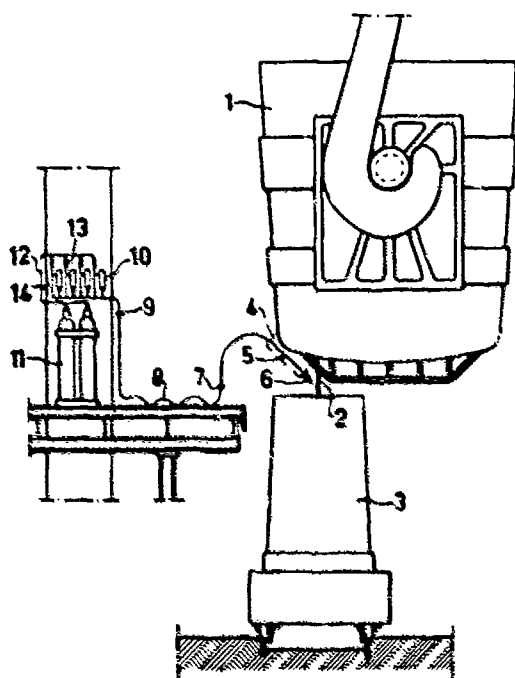


FIG. 8. ARRANGEMENT FOR INJECTION INTO THE MOLD OF A COLUMBIUM CONTAINING POWDER. NORMALLY FERRO-COLUMBIUM. THE POWDER IS BLOWN INTO THE STEEL STREAM ON CASTING AN INGOT. FROM NITROGEN OR ARGON CONTAINERS (11) THE GAS WILL UNDER PRESSURE BLOW THE POWDER IN THE FERRO-COLUMBIUM CONTAINERS (12) INTO A SMALL DIAMETER STEEL TUBE (5), WHICH IS APPLIED ON THE LADLE SO THAT THE POWDER STREAM WILL ALWAYS BE DIRECTED ON THE STEEL STREAM DURING THE CASTING.

Besides the higher recovery of the micro-alloying element used, this method has another advantage--the distribution of small additions will be more even through an ingot series than by any other method we have investigated.

This has been shown by means of radioactive isotopes and, of course, also by means of more conventional investigation methods such as analyzing different parts of steel plates and testing the mechanical properties of the plates accordingly.

Another advantage connected with addition to the mold is that in case of great heat weights only a selected part of the ingots have to be produced as micro-alloy steels, while the rest of the charge may be used for other purposes.

On using this principle for producing ingots of the same basic heat with various additions of micro-alloying metals, the composition of the steel in the ladle should of course correspond to an ordinary structural steel, a carbon steel or a carbon-manganese steel, i.e. a ship steel. Since the columbium addition to a steel calls for certain composition limits of the elements in the base composition of the steel in the ladle, there will always be a possibility to stop the addition of the micro-alloying element in case the composition requirements of the base steel have not been met when the steel has been tapped into the ladle. If so, the base steel can still be used provided that it corresponds to the requirements of an ordinary structural steel and will then be used for ingots of this type.

Our investigations have also covered studies of different ingot sizes and types. We have found no significant difference concerning the properties of the columbium steel, which can be said to depend on these factors.

It might further be of interest that columbium-containing scrap will give off its columbium contents to the slag on re-melting. The process of oxidation seems to run to a very low content of columbium in the molten steel, the distribution ratio of columbium in slag/columbium in steel being 200-300/1 under strongly oxidizing conditions. On re-circulating columbium-containing slag to the blast furnace, however, attention should be given to the possibility of gradually increasing columbium-contents in the pig iron caused by reduction of columbium-oxide.

Our experience of mechanical treatment is limited to slab and plate rolling. The slab rolling has been performed without any particular precautions. The heating temperature before rolling is normally 1240-1300 °C. The only

trouble that has occurred in connection with the slab rolling is that on rather high columbium additions, resulting in columbium contents of the steel in the order of 0.04-0.05%, the slabs may become rather brittle. On surface conditioning of such slabs, they have in some cases broken in two due to brittle fracture initiated at some defect in the slab under the influence of the thermal stresses. Such incidents are of course exceptions.

The heating before plate rolling is normally carried out at a temperature of about 1200°C and the plate rolling is performed under controlled temperature conditions.<sup>12</sup> These conditions normally imply 30% reduction at a temperature below 900°C.

A great many other variables of hot rolling conditions have been investigated. There does not, however, seem to be any need for further restrictions, but on the other hand, the amount of reduction below 900°C mentioned above has to be fulfilled to ensure the desired properties of the plates.

The lamination tendency of columbium steel plates does not seem to be stronger than for ordinary carbon steels or carbon-manganese steels. On the other hand, there is a difference between a columbium-treated steel and an aluminum-treated one. Columbium has definitely not the same marked effect on the slag distribution and the ferrite banding of the microstructure as has aluminum.

The same practice as to ultrasonic testing of normal structural steel plates can be applied to the columbium steels.

Plain carbon steels or carbon-manganese steels with micro-alloying additions of columbium are delivered either in the hot-rolled condition up to a certain plate thickness or after normalizing.

The normalizing treatment does not distinctly differ from the same treatment of plain carbon steels or carbon-manganese steels. The normalizing temperature is about 900°C but too low a normalizing temperature seems to be more detrimental for columbium steels than for unalloyed or manganese alloy materials.

The heat treatment practice to be chosen is in most cases practically unchanged by very small additions of alloying elements. There are exceptions, of course, for instance in:

connection with boron additions to certain steel types, but generally most of the micro-alloying elements will not change the heat treatment conditions of the base steel more than some  $\pm 10^\circ\text{C}$  with regard to the  $A_1$ -level.

#### BASIC PROPERTIES OF COLUMBIUM STEELS VERSUS PROCESSING VARIABLES

The properties of columbium steels described in this part of the report refer mainly to the three compositions given in Table 1 and to surrounding compositions investigated in our research work.

##### Properties vs. Composition

A normal microstructure of a columbium steel (steel A, Table 1) in the hot-rolled condition at a plate thickness of 30 mm is shown in Fig. 9. There is not very much difference in microstructure at still higher plate thicknesses.

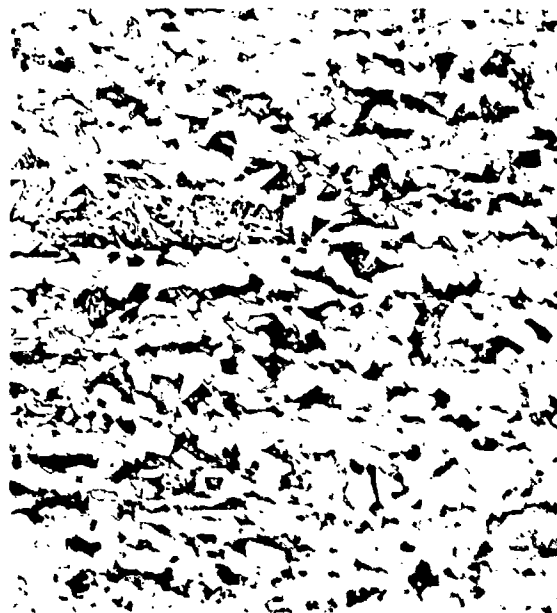


FIG. 9. MICRO-STRUCTURE OF A STEEL WITH THE COMPOSITION C 0.21%, Si 0.11%, Mn 1.39%, Cb 0.056% IN THE AS-ROLLED CONDITION. GRAIN-SIZE ASTM 8. 200 x

The variations in microstructure by various columbium additions outside the range 0.005-0.05% has not yet been properly investigated in our research work. Within the mentioned range, however, no great variations have been observed until now with the exception of some-

what increased grain-refinement with increasing columbium contents.

The microslag types to be found in columbium-treated steels are shown in Fig. 10. As



FIG. 10. MICRO STRUCTURE OF A STEEL WITH THE COMPOSITION C 0.11%, Si 0.02%, Mn 1.01%, Cb 0.036%. IN THE AS-ROLLED CONDITION, SHOWING TYPICAL SILICATE SLAG INCLUSIONS. GRAIN-SIZE ASTM 7-8. PLATE THICKNESS 30 mm. 400 x

in ordinary structural steels, the slag inclusions are of the sulphide type and the silicate type. By means of electron probe X-ray micro-analysis, however, we have found that the silicate inclusions may contain up to 3% columbium. This might call for a deoxidation practice that will guarantee the smallest possible amount of oxygen-containing slag inclusions.

The distribution of columbium carbides in the structure, is difficult to observe directly in a normal light-microscope. However, at medium magnifications and proper etching, it is sometimes possible to observe particles that are probably such carbides, and particularly so after having studied these precipitations in the electron microscope. Following this approach, they are more easily found in normal microscopy.

The distribution of columbium in the microstructure can further be investigated in a more

adequate way by means of the electron probe X-ray micro-analysis. Such investigations have proved that columbium is not evenly distributed in the microstructure after rolling. A columbium concentration will always be found at the grain boundaries, and a higher columbium content in the pearlite than in the ferrite has also been observed.

The columbium contents of the grain boundary areas are normally about three times as high as in the ferrite. It is still not clear whether this distribution of columbium has any importance as to the properties of the steel or if it can be influenced by hot-rolling conditions or any other processing variables, etc. It has been observed, however, that normalizing will result in a more even distribution.

It seems most probable that the columbium distribution is quite important. However, this is a part of the research field still including many unknowns and calls for further investigations. What we dare say today is, by our experiences, that on keeping the same processing conditions from time to time, the columbium distribution will be found to be the same in each case, i.e. the distribution of columbium is probably strongly connected with the treatment of the steel and will not vary independently of this from one heat to another. The carbide distribution after normalizing is evidently a reasonable explanation of the corresponding change in properties.

The  $A_c$ -temperature is very slightly increased by columbium contents in the order of 0.02-0.04%. Our investigations have shown that such a columbium addition will raise this critical temperature about 10°C and that a further addition of the same amount of vanadium will increase  $A_c$  another 5°C. From a practical point of view these changes have no importance.

An increase of the columbium contents will cause an increased stability against spontaneous grain growth. On overheating, the temperature of sudden grain growth will be found around 1000°C, if 15% coarse-grained structure is taken as the criterion. Killed columbium steels possess a higher grain-growth temperature and the difference between a killed steel and a semikilled one is about 50°C. The grain-growth tendency of columbium steels is less drastic than for aluminum-treated steels of the same basic composition (Fig. 11-13).

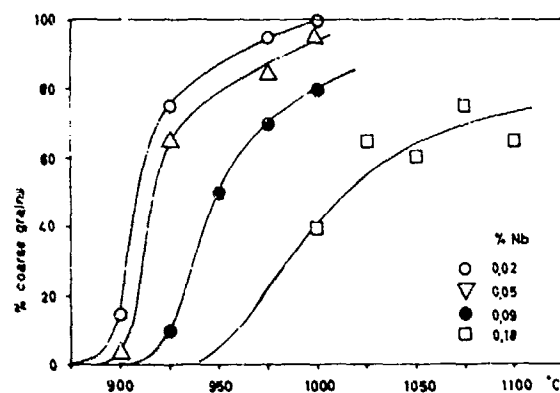


FIG. 11. INFLUENCE OF VARIOUS COLUMBIUM CONTENTS ON GRAIN GROWTH TENDENCY OF A STEEL WITH THE BASIC COMPOSITION C 0.17%, Si 0.09%, Mn 0.43% AND COLUMBIUM (NIOBIUM) CONTENTS ACCORDING TO THE DIAGRAM.

Columbium additions to a steel will increase the yield strength, the ultimate strength and the ratio yield strength/ultimate strength.

Up to columbium contents of about 0.02%, the influence on the properties mentioned occurs very strongly. A further increase above this columbium level will still slowly raise the yield strength, while the ultimate strength does not seem to be markedly influenced. This is the case up to about 0.10% columbium, while further additions up to about 0.20% columbium will cause a continuous slight decrease in ultimate strength--the  $\sigma/\sigma_u$  remaining almost unchanged.

Our investigations have indicated that, as an example, the yield strength/ultimate strength ratio, which for a certain carbon-manganese steel is about 0.60, will increase up to about 0.68 at an addition of 0.004% columbium, up to 0.73 at 0.01% columbium, but only up to 0.77 at a further addition up to 0.06% columbium.

Even if the strongest influence of columbium on the yield strength apparently occurs already at contents of the order of 0.01% it seems from a practical point of view to be reasonable to add an average content of 0.02-0.03% in order to avoid a detrimental influence of unavoidable segregations on full-scale ingot production.

Apart from this direct influence of columbium on yield strength and ultimate strength,

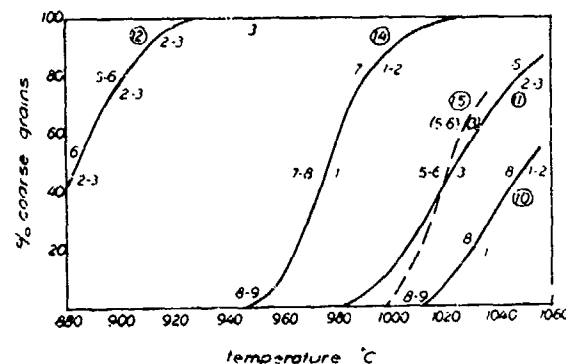


FIG. 12. GRAIN GROWTH TENDENCY OF VARIOUS HEATS WITH THE FOLLOWING COMPOSITIONS (HEAT NUMBER INDICATED IN THE DIAGRAM BY THE FIGURES WITHIN THE CIRCLES):

NO.	C	Si	Mn	P	S	Co
10	.10	.27	1.12	.043	.046	0.24
11	.13	.36	0.89	.041	.046	0.11
12	.19	.07	1.17	.058	.050	0.00
14	.19	.12	0.86	.067	.044	0.10
15	.16	.29	1.09	.064	.045	0.11

THE OTHER FIGURES ALONG THE GRAIN GROWTH CURVES FOR THE DIFFERENT STEELS INDICATE THE GRAIN SIZE NUMBER ACCORDING TO ASTM. IT SHOULD BE OBSERVED THAT AT HIGHER TEMPERATURES THE MICRO-STRUCTURE CONSISTS OF A MIXTURE OF FINE AND COARSE GRAINS AS SHOWN BY THE ASTM NUMBERS ON EACH SIDE OF A CURVE.

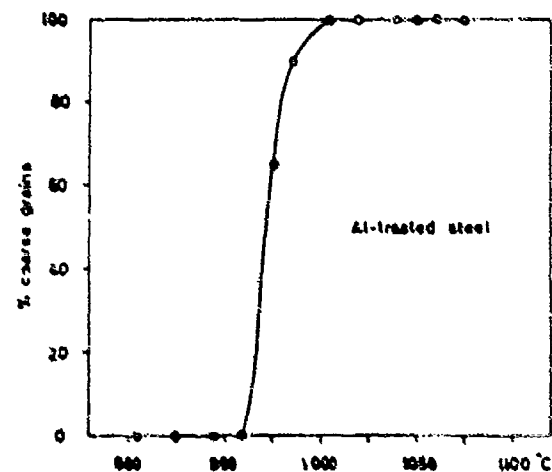


FIG. 13. DIAGRAM SHOWING SPONTANEOUS GRAIN GROWTH IN AN ALUMINUM-TREATED CARBON-MANGANESE STEEL WITH C 0.15%, Si 0.25%, Mn 0.95%.

any change of the basic composition of the steel will cause a corresponding change in strength, which means that in case of constant columbium contents the strength of the steel may be changed in a normal way by changing the carbon contents, the manganese contents, etc. The influence of columbium is, in other words, to be regarded as one which is added on the top of the normal strength of the base alloy.

As a consequence of increasing the yield strength by columbium additions there is a corresponding tendency to decrease the elongation, which, however, does not seem to be critical within rather wide limits.

There is also a change in impact properties to be observed, following the increase of yield strength. In most of the literature references to be found concerning the influence of columbium on steel it is claimed that the impact properties of columbium steels are good and in many cases improved in relation to columbium-free steels. I think I dare say that our investigations have covered enough impact studies to state that this is definitely not true regarding columbium-treated plain carbon steels and carbon-manganese steels in the hot-rolled condition. It might be true regarding some cases of normalized or quenched-and-tempered conditions and it is definitely true concerning columbium-treated low-alloy hardened and tempered steels. In the latter case, however, it occurs as a consequence of columbium additions in the order of 0.20-0.40%.

As to the columbium-treated carbon steels and the corresponding carbon-manganese steels this statement does not mean that the impact properties are very poor. I only claim that an improvement hardly occurs because of a columbium addition only and already this statement might be an understatement.

#### Properties vs. Rolling Conditions

The influence of rolling conditions on the microstructure of columbium steels is much the same as on aluminum-treated steels. On controlled rolling a more fine-grained structure will form and a certain tendency to ferrite banding may accordingly appear. This ferrite banding, however, is not much pronounced even if the finishing temperature on rolling is lowered very much. In this respect the difference between columbium steels and aluminum-treated steels is obvious.

The strongest influence of rolling conditions will be found on the mechanical properties and, particularly, with regard to the impact values of the steel.

Our research work has covered a great many variables in connection with hot rolling. No significant effects have been observed as to reasonable changes in heating temperatures before rolling, various cooling rates immediately after rolling or various temperatures on levelling the plates after rolling. Nor have more complicated prescriptions for controlled rolling resulted in properties, which deviate from the properties gained by a normal controlled rolling, i.e. a certain reduction below a certain temperature. Variations within a wider heating range, e.g. 100°C, before rolling, however, will result in rather strong effects on mechanical properties.

Regarding the ultimate strength of a columbium steel the finishing temperature on hot rolling has only a very small influence and, as a consequence of what has been said above concerning cooling rates after hot-rolling, etc., the influence of plate thickness on the ultimate strength is for the same reason limited if, on the whole, it can be observed.

The yield strength, however, is more obviously influenced by the finishing temperature on rolling and also by the degree of reduction below a certain temperature.

We have found that there seems to be an optimum concerning the impact strength level around a finishing temperature of 830°C.

It can also be shown that the ratio yield strength/ultimate strength will increase on increasing reduction below 900°C. Hence this ratio will cover the range 0.74-0.78 by increasing the degree of reduction below 900°C from 30% to 70%, as far as our investigations have shown. For normal hot rolling, i.e. without attention to any controlled conditions, the same ratio will be in the order of 0.67-0.72 depending on plate thickness.

In other words, the finishing temperature will have roughly the same influence on columbium-treated steels as on columbium-free steels of the same basic composition although we feel that the columbium-free steels may show a little stronger effect by varying hot-rolling conditions than do the columbium-



treated ones.

The particular effect of the columbium addition, on the other hand, will naturally cause higher absolute values of yield strength/ultimate strength over the whole line of hot-rolling variables.

The elongation is obviously strongly related to the strength of the steel. Elongation values could only be compared provided the strength in various cases is about the same.

Approximately ultimate strength x elongation is constant. This is a rather well-known expression but it is unclear within which range it is valid. In our investigations we have used the expression yield strength x elongation, which at constant yield strength/ultimate strength will imply the same as the previously mentioned one.

We have called yield strength x elongation ( $\text{kg/mm}^2 \times \delta, \%$ ) the Q-value. This Q-value will normally vary between 960 and 1190 with an average value of 1080 if calculated on the basis of our investigation results. Within a certain heat, however, the scattering is less than the range mentioned.

The reason why, on the whole, the Q-value will vary is for the present unknown to us. A Q-value of minimum 1000 is for most purposes demanded in our production as a reasonable relationship between yield strength and elongation.

For a columbium-treated steel the Q-value is higher than for a corresponding columbium-free steel, while ultimate strength x elongation is somewhat lower for the columbium steel.

For a certain yield strength, columbium-treated steels have a better elongation than corresponding columbium-free steels and vice-versa if the ultimate strength is kept constant.

No relation between ultimate strength x elongation and hot-rolling conditions (including heating conditions before rolling) has been found, although such a relationship might exist between the Q-value and the rolling conditions. This is still being investigated.

The impact properties of a columbium steel are strongly depending on hot-rolling conditions. Our investigations have given a great

many results concerning the variation of impact resistance with respect to controlled-rolling conditions used. They can be summarized as follows:

1. The impact resistance of a columbium steel in the hot-rolled condition is, whatever the rolling conditions may have been, inferior to an unalloyed or manganese-alloy steel of corresponding basic composition.
2. The impact resistance of a columbian steel is strongly influenced by decreasing finishing temperature on hot-rolling and by increasing the degree of reduction below the control temperature chosen.
3. The effect of lowering the finishing hot-rolling temperature is very pronounced down to 800°C. A further temperature decrease will not, however, lead to a corresponding improvement of the impact resistance. In principle the same is true down to a certain degree of reduction below the finishing temperature. Our investigations have shown that 30% reduction below 900°C will give a marked effect while further increase of the reduction below the chosen control temperature will not lead to a corresponding improvement.
4. Besides the influence of hot-rolling conditions, the impact resistance of a columbium steel is, of course, also dependent on the steel composition and further on the deoxidation practice. Hence the impact strength of a silicon-killed columbium steel is better than that of a semikilled steel, but still inferior to that of a corresponding aluminum-treated one. Between these three steel types the difference in lower transition temperatures on Charpy V-notch testing is about 10°C.

Concerning the standard deviation of various basic mechanical properties of columbium steels with reference to a continuous production of this steel type and the standard deviation of a certain heat of a columbium steel respectively, there does not seem to be a more pronounced one than for ordinary structural steels. In other words scatter readings on mechanical testing of columbium steels have not been found to be caused by the columbium addition as such but is rather a consequence of variations caused by the basic composition.

There is a systematic decrease in yield strength from the top end of the ingots to the

bottom end, but this is also true for ordinary structural steel ingots.

#### Properties vs. Heat Treatment

A great many investigations concerning heat-treatment conditions in connection with normalizing have been performed (Fig. 14).

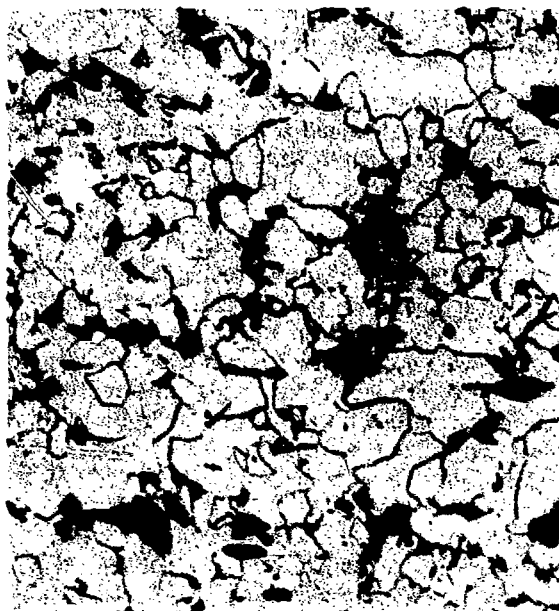


FIG. 14. MICRO-STRUCTURE OF THE SAME STEEL AS IN FIG. 10 AFTER NORMALIZING. IT IS TYPICAL FOR A COLUMBIUM MICRO-ALLOY STEEL THAT THE GRAIN SIZE IN THE NORMALIZED CONDITION VARIES CONSIDERABLY. GRAIN-SIZE ASTM 8-10. 400x

The  $A_1$ -temperature of columbium steels discussed in this paper is 840-850°C. The normalizing temperature is generally 900-920°C. Without going into details the heat-treatment investigations can be summarized as follows:

1. Normalizing will decrease the ultimate strength to a level, which is 1-2 kg/mm<sup>2</sup> (1492-2800 psi) higher than the ultimate strength of a corresponding columbium-free steel, independently of hot-rolling conditions.
2. Normalizing will reduce yield strength/ultimate strength ratio to about 0.70.
3. The impact properties of normalized columbium steels will increase in relation to the same steel in the hot-rolled condition and will in most cases become quite comparable with or

superior to the impact values of a corresponding columbium-free, normalized steel or even a normalized aluminum-treated one.

4. Normalizing will improve the Q-value but not very strongly.

In connection with the heat-treatment investigations the properties of columbium steels after hardening and subsequent tempering have also been studied. This part of our investigations has, however, until now covered only a small part of what we intend to do and it might be a little early to draw any conclusions. A columbium-carbide precipitation with its maximum around 550-600°C can, however, be reported (Fig. 15) after solution treatment at sufficiently high temperature, e.g. 1250°C.

The susceptibility to aging is generally less for a columbium steel as compared with a corresponding columbium-free steel.

In most cases columbium steels are, in the normalized condition, as good as corresponding aluminum steels, and a good deal of our recent investigations have proved the aging tendency of columbium steels to be less pronounced than that of normal fine-grained aluminum steels.

#### THREE POSTULATES

1. The function stability of a welded structure depends on the frequency and types of defects in the welded joints.\*
2. It is unrealistic to believe that a welded structure of any importance is completely free from defects in its welded joints.
3. All precautions taken in connection with welding have the aim to decrease, in one way or another, the level of welding stresses and/or to prevent the occurrence of injurious micro-structure formations which may increase a dangerous influence of appearing defects.

I am quite convinced that these postulates are valid. If so, the consequence will be that every catastrophic service failure, known to have occurred in a welded structure, must

\* Weld metal + surrounding heat-affected zones and parts under the influence of welding stresses.

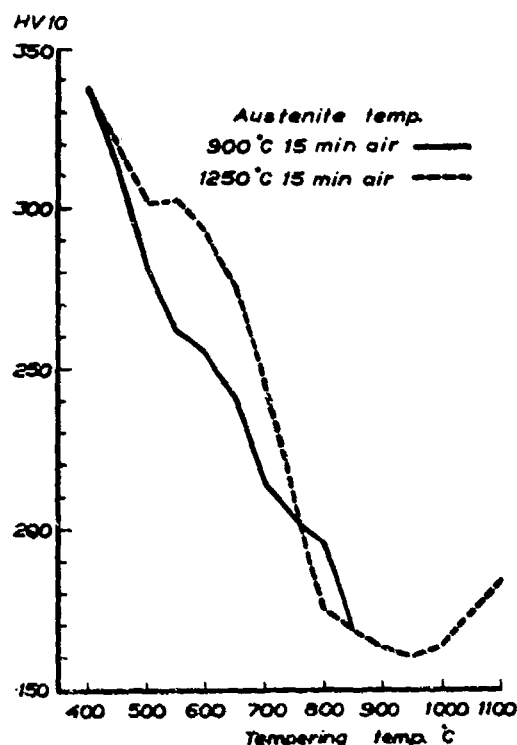


FIG. 15. DIAGRAM SHOWING HARDNESS VS TEMPERING TEMPERATURE OF STEEL C IN TABLE I AFTER AIR-COOLING FROM 900°C AND 1250°C RESPECTIVELY. AFTER SOLUTION TREATMENT AT SUFFICIENTLY HIGH TEMPERATURE A MARKED COLUMBIUM CARBIDE PRECIPITATION WILL APPEAR AFTER TEMPERING AROUND 550-600°C.

have originated from a defect in a welded joint under circumstances which have been critical with respect to temperature and stress conditions.

So far as I know nobody has been able to prove that this is not true.

In the following part of this report some special properties of columbium steels in connection with welding technology will be described. Before that the seven groups of factors to be observed in connection with weldability investigations as to detrimental changes should be repeated:

1. Longitudinal weld cracking (solidification cracks or "hot cracks").
2. Transversal weld cracking (shrinkage cracks or "cooling cracks").

3. Hardening embrittlement in the weld or the transformation zone of the steel.

4. Normal brittle behavior of the weld or the steel below a characteristic critical temperature and under severe stress conditions, e.g. residual welding stresses.

5. Embrittlement due to microstructure instability of the weld or the steel at low and medium temperatures.

6. Embrittlement due to microstructure instability of the weld or the steel at high temperatures.

7. Decrease of corrosion and oxidation resistance of the weld or the steel due to residual welding stresses and/or certain microstructure formations.

#### SPECIAL PROPERTIES OF COLUMBIUM STEELS VS. WELDING TECHNOLOGY

Having now described, in a summarized form, the basic properties of columbium-alloy steels vs. processing variables I ought to turn back to the weldability problems connected with this type of steel and describe how columbium-alloy structural steels will react and should be regarded in connection with welding technology.

Welding technology does not only include what is generally called weldability problems but also problems caused by cutting and forming operations, choice of filler materials and determination of suitable preheating or post-heating temperatures, if such precautions are necessary in certain cases. A successful handling of these problems and avoiding the detrimental effects, which may arise from the metallurgical reactions during the welding, is a requirement to be fulfilled in order to offer a high degree of function stability to a welded joint in the steel.

Concerning columbium steels of the type discussed here, one will not meet any particular problems, so far as I know, with regard to cutting, forming operations and choice of filler materials.

There is a difference, of course, between plain carbon steels or carbon-manganese steels and the columbium-alloy steels respectively

caused by the higher strength of the latter. But filler materials, which must not necessarily or not even preferably be columbium-alloy materials, can easily be found as they have corresponding strength properties.

Provided the equipment used for forming can be applied to steels with higher strength, difficulties which may arise are of the same type as will occur for unalloyed or manganese-alloy steels. During our investigations no serious troubles have appeared according to the factors mentioned, which could not have occurred in columbium-free steels as well.

It has previously been mentioned that there are seven main groups of metallurgical phenomena to be particularly studied in connection with weldability investigations and that certain detrimental changes may be expected under circumstances as a consequence of these metallurgical reactions.

Columbium as a micro-alloying element in a structural steel does not seem to contribute to either longitudinal or transversal weld cracking. In these respects a columbium-alloy steel will react as a corresponding plain carbon or carbon-manganese steel.

For example, the main reason for hot-cracking in welds is too high carbon contents and/or sulphur contents. Neither an advantage, nor a disadvantage of a small columbium addition has been found.

In the same way transversal weld cracking is a problem connected with the weld metal quality and the shrinkage-stress conditions during welding. Small columbium additions to the weld metal from the molten steel does not seem to have any practical importance.

There is no obvious reason to expect that columbium in the steel will contribute to a decrease of the corrosion and oxidation resistance of the parts of the base metal surrounding welds in such a steel. This is, on the other hand, a part of the weldability research, which has not yet been investigated in our work.

More interesting parts of our weldability investigations refer to the risk of hardening embrittlement in the transformation zones of a columbium steel adjacent to a weld, to the risk of initiation of brittle failures in or a-

round welded joints in columbium steels and to the possible change in properties, which such a steel may undergo because of thermal instability during heating to medium or high temperatures during or after welding.

On rapid heating and cooling, as during welding, a rather pronounced effect of columbium can be observed. This can be shown by means of a special weld-hardening test based upon high-frequency induction heating of test bars, whereby the heating and cooling cycles on welding can be reproduced.<sup>1</sup>

Since there is no welding included in this type of hardenability testing, which is briefly described in Appendix A, the testing conditions are from time to time kept very strictly.

It is well-known that hardenability diagrams as they appear on Jominy testing have been used for quite a few years in order to determine welding conditions for various steel types. The induction-heated weld-hardening test mentioned will offer a hardenability curve for the steel, which can be used in the same way and which has been developed so that the same tables as for the Jominy hardenability diagrams can be used for calculations of welding conditions--but with the important difference that the heating and cooling conditions on welding can be simulated in a far better way.

Provided that the steel to be tested does not contain any alloying elements forming carbides, which very slowly will be brought into solution on austenitization the Jominy test could be used as well. However, as soon as slowly dissolving microstructural elements occur, such as carbides of strong carbide formers, the Jominy curve will not offer a true picture of the hardenability of a heat-affected zone close to a weld.

The Induction-heated weld-hardening test, which was developed about ten years ago, has proved to be very useful for the determination of slight differences in hardenability of various structural steels. Figure 16 shows a hardenability curve received by the Induction-hardening test. This curve should be compared with the curve in Fig. 17 for a corresponding carbon-manganese steel without columbium addition.

It is evident that an advantage has been

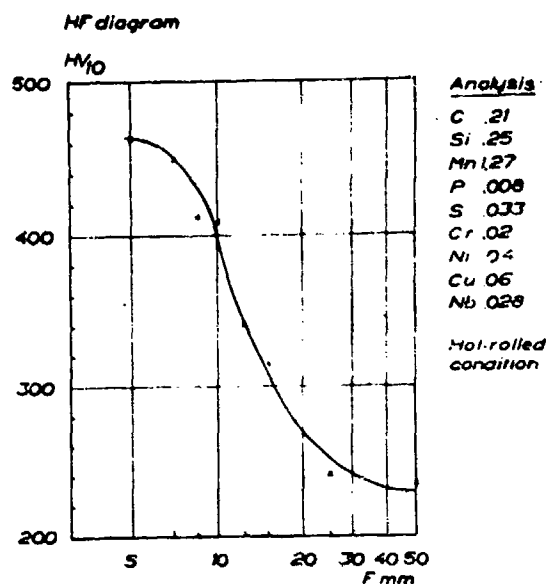


FIG. 16. NWH HARDENABILITY DIAGRAM FOR A CARBON-MANGANESE STEEL WITH A COLUMBIUM MICRO-ALLOY ADDITION. IN SPITE OF A YIELD STRENGTH, WHICH IS 25-30% HIGHER THAN THAT OF THE CARBON-MANGANESE STEEL IN THE NEXT FIGURE, THE HARDENABILITY IN CONNECTION WITH WELDING CAN BE KEPT MUCH LESS BECAUSE OF LOWER MANGANESE CONTENTS. SEE ALSO THE DIAGRAM IN APPENDIX C, WHICH SHOWS THE ADVANTAGE OF CHOOSING A HIGH STRENGTH MICRO-ALLOY STEEL INSTEAD OF A NORMAL CARBON-MANGANESE STEEL WITH RESPECT TO THE HARDENING RISK IN THE HEAT-AFFECTED ZONES ADJACENT TO WELDS.

gained by alloying the steel with a small amount of columbium if a rapid heating is applied. Consequently columbium steels must in this respect be regarded as having improved weldability in relation to their strength.

The slow rate by which columbium carbides may go into solution in the austenite does not occur to me as a probable explanation of this behavior of a columbium steel. Still the austenite in a heat-affected zone in a columbium steel is probably lower in carbon than in the steel according to the actual composition and the critical cooling rate of such an austenite will become higher. However, the columbium contents are indeed not sufficient to form any appreciable amount of columbium carbides. I feel that columbium rather may form an essential part of the cementite but we have not been able to prove this, yet. On the other hand, the

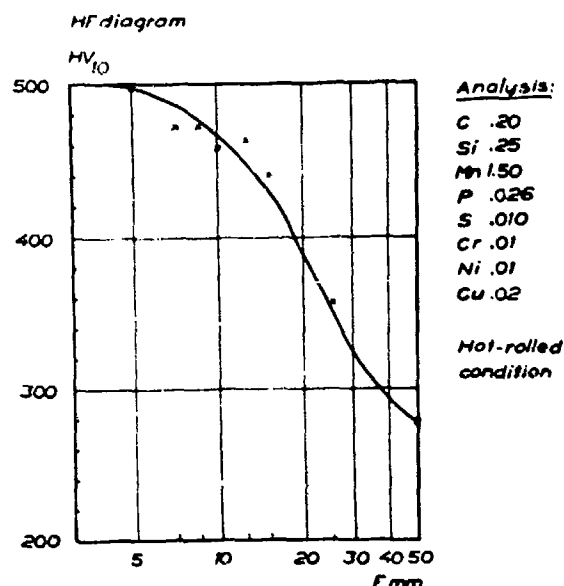


FIG. 17. NWH HARDENABILITY DIAGRAM FOR A CARBON-MANGANESE STEEL WITH CARBON AND MANGANESE CONTENTS CLOSE TO THE UPPER LIMIT OF WHAT IS NORMALLY PERMITTED WITH RESPECT TO WELDABILITY AND STILL A MUCH LOWER YIELD STRENGTH THAN THE STEEL REPRESENTED IN THE FOREGOING DIAGRAM.

columbium influence on the cementite formation and localization is pronounced. The pearlite will precipitate in an abnormal shape; the cementite appears to a certain extent in the grain boundaries and is rather coarse. Finally, by means of X-ray probe micro-analysis it has been shown that the ratio of columbium contents in the grain boundaries, in the pearlite and in the ferrite are in the relative amounts of about 3-1 1/2-1.

The tendency to brittle fracture in a columbium steel in connection with welding will offer much of interest. It can be shown and has already been said previously in this paper that the impact strength vs. temperature of a columbium steel is generally not better and, as rolled, rather worse than what can be expected regarding a corresponding steel without columbium. Hence one could easily be tempted to state, from this point of view, that columbium steels are normally inferior to the corresponding unalloyed or manganese-alloy ones.

This impression is no doubt obtained if the brittle-fracture tendency is determined only by means of impact testing of the unwelded steel.

The question is however, if it is correct to exclude welding from such a testing.

In most cases it is done so because nobody, so far as I know, has been able to show that any remarkable improvements can be gained by the heat-influence of welding as to the safety against brittle fracture of a steel. It is rather a rule (or at least believed to be a rule) that the heat-affected parts of a base metal are inferior to the unaffected steel in this respect.

Figures 18-19 show quite normal impact curves of a columbium steel in the hot-rolled and normalized conditions respectively. There is nothing abnormal in the curve referring to the normalized condition and in this case the steel

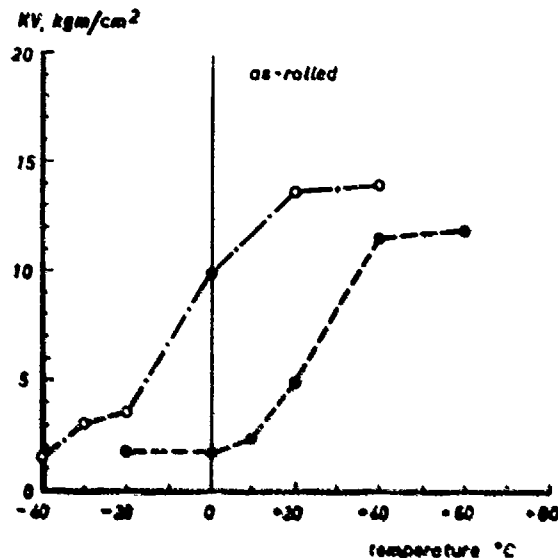


FIG. 18. CHARPY V-NOTCH IMPACT CURVES FOR 30 mm STEEL C IN TABLE 1 IN THE AS-ROLLED CONDITION (BLACK DOTS) AND FOR A CORRESPONDING CARBON-MANGANESE STEEL WITHOUT COLUMBIUM ADDITION BUT STILL IN THE AS-ROLLED CONDITION AND WITH THE SAME PLATE THICKNESS.

has a good chance to withstand severe stress conditions caused by sharp notches even at rather low temperatures. The hot-rolled condition of the steel, however, does not create any happy feelings even if there are lots of columbium-free steels with roughly the same brittle fracture tendency already at high temperatures.

It has previously been said a few words about the aging susceptibility of columbium steels.

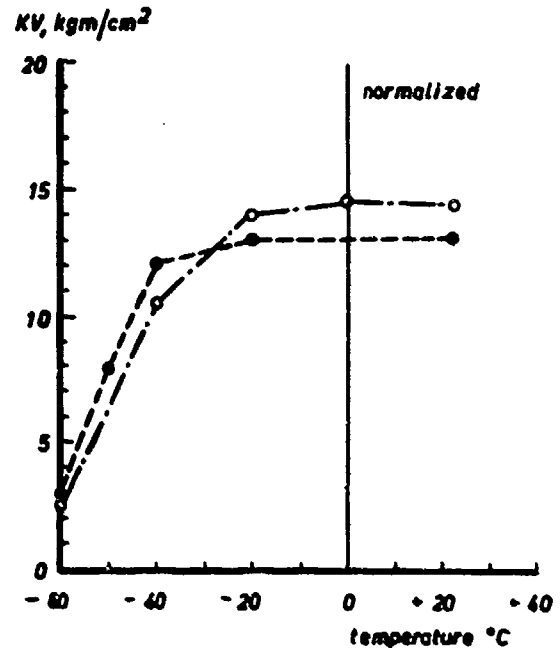


FIG. 19. THE SAME STEEL PLATES AS IN FIG. 18 AFTER NORMALIZING.

The aging reaction in a columbium steel may occur already on rapid heating in connection with plastic deformation as, for instance, in a zone at a certain distance from a weld. A decreasing impact strength in such parts of a columbium-alloy base metal is shown in Fig. 20. This is neither worse nor better than what is to be found for most structural steels.

However, on testing a columbium steel by the NC-testing method (Appendix B) quite another picture of the brittle fracture tendency will appear. This is shown in the diagrams of Fig. 21-22. The two diagrams represent examples of the hot-rolled and the normalized condition respectively. It can be seen that the transition temperature is very low indeed, in both cases about  $-100^{\circ}\text{C}$ .

In spite of the sharp notch attack from a natural weld crack it has been impossible to cause a fracture in a test bar above  $-100^{\circ}\text{C}$  at nominal loads below the yield strength level of the steel, i.e. the nominal yield strength measured on unnotched test bars. This is the same yield strength level as measured by means of welded test bars above the intersection point between the yield strength curve and the curve of the so-called nominal cleavage strength.

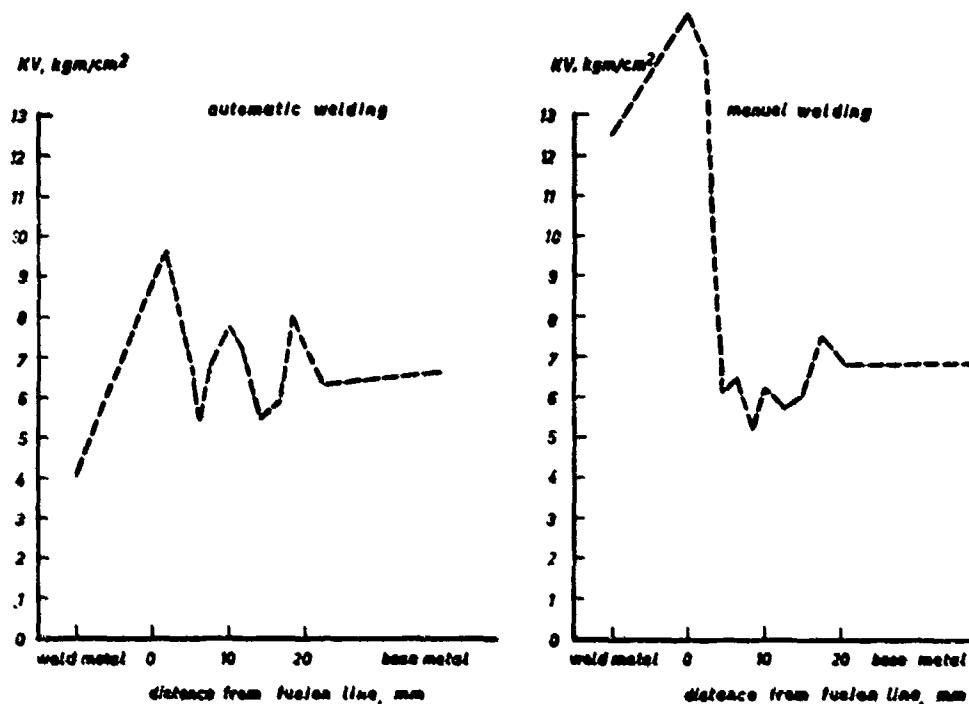


FIG. 20. IMPACT STRENGTH ACCORDING TO CHARPY V-NOTCH TESTING AT 0°C IN AND AROUND A WELDED JOINT IN A HALF INCH STEEL OF THE TYPE B IN TABLE 1 (AUTOMATIC WELDING LEFT, MANUAL WELDING RIGHT). THE STEEL WAS IN THE AS-ROLLED CONDITION AND A SLIGHT DECREASE IN IMPACT STRENGTH CAN BE OBSERVED AT A DISTANCE OF 5-15 mm FROM THE FUSION LINE. THE TWO MINIMA WITH A MAXIMUM IN IMPACT STRENGTH IN BETWEEN ON AUTOMATIC WELDING AS WELL AS ON MANUAL WELDING HAVE BEEN REPRODUCED FREQUENTLY BUT ARE NOT YET FULLY EXPLAINED. HOWEVER, THE MINIMUM AT A DISTANCE BETWEEN 5-10 mm FROM THE FUSION LINE IS PROBABLY CAUSED BY AN AGING REACTION BUT THE OTHER ONE MAY HAVE ANOTHER REASON. IT SHOULD FURTHER BE NOTED THAT AS EXPECTED THE IMPACT STRENGTH OF THE WELD METAL OF THE AUTOMATIC WELD IS FAR LOWER THAN IN THE MANUAL WELD. BUT FURTHER THAT A PRONOUNCED MAXIMUM IN IMPACT STRENGTH APPEARS CLOSE TO THE FUSION LINE IN BOTH CASES. THIS MAXIMUM IS ALSO REPRODUCIBLE AND WILL BE EXPLAINED BY THE FIG. 21-25. THE DIAGRAMS ARE TAKEN FROM AN UNPUBLISHED INVESTIGATION BY B. RAMSHAGE, A CO-WORKER OF THE AUTHOR.

It will further be observed that at very low temperatures, about -200°C, the nominal cleavage strength of this columbium steel (as well as of other corresponding columbium-alloy steels) is still surprisingly high, about 20 kg/mm<sup>2</sup> (28,500 psi). At this temperature one will find the intersection between the conventional curves for ultimate strength and yield strength. Consequently, and according to the interpretation of the NC-testing results, this will simply imply that the stress level necessary for the propagation of an initiated brittle fracture is of the same order of

magnitude, i.e. about 20 kg/mm<sup>2</sup>. This is higher than for most other structural steels and definitely higher for columbium steels than for any other structural steel with a corresponding yield strength as far as our investigations have shown.

However, a great many NC-testing investigations have been performed and in all cases there is a very good relationship between the temperature for the intersection of the NC-curve and the nominal yield strength curve on one side and the critical impact level on Charpy V-notch testing on the other. The lat-

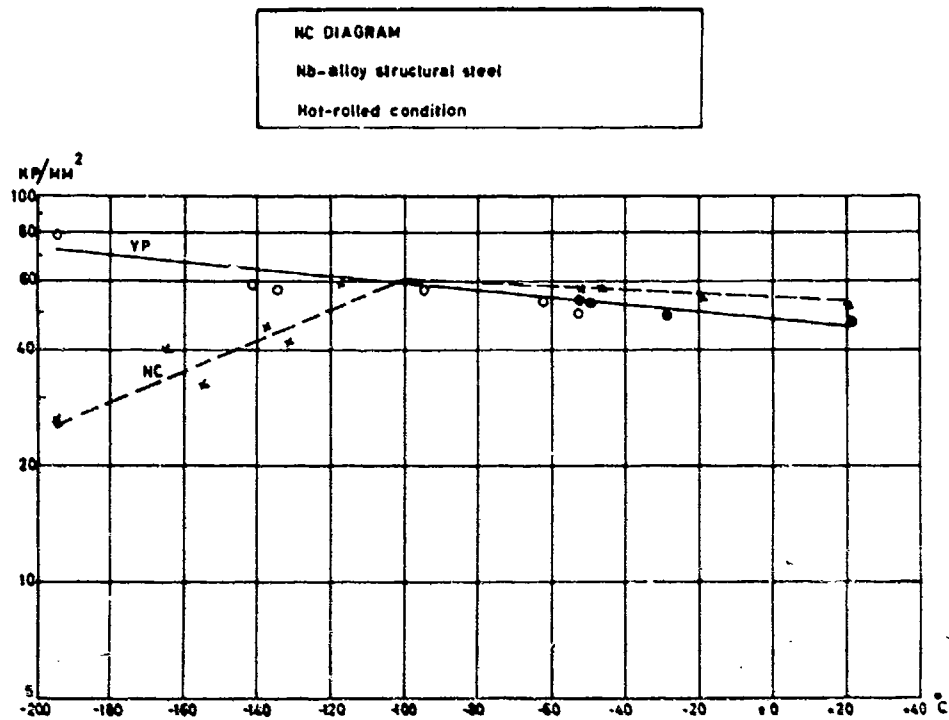


FIG. 21. NC-DIAGRAM FOR 20 mm STEEL C IN TABLE 1 IN THE AS-ROLLED CONDITION. THE TRANSITION TEMPERATURE  $T_c$  (SEE APP. B) IS ABOUT  $-100^{\circ}\text{C}$ .

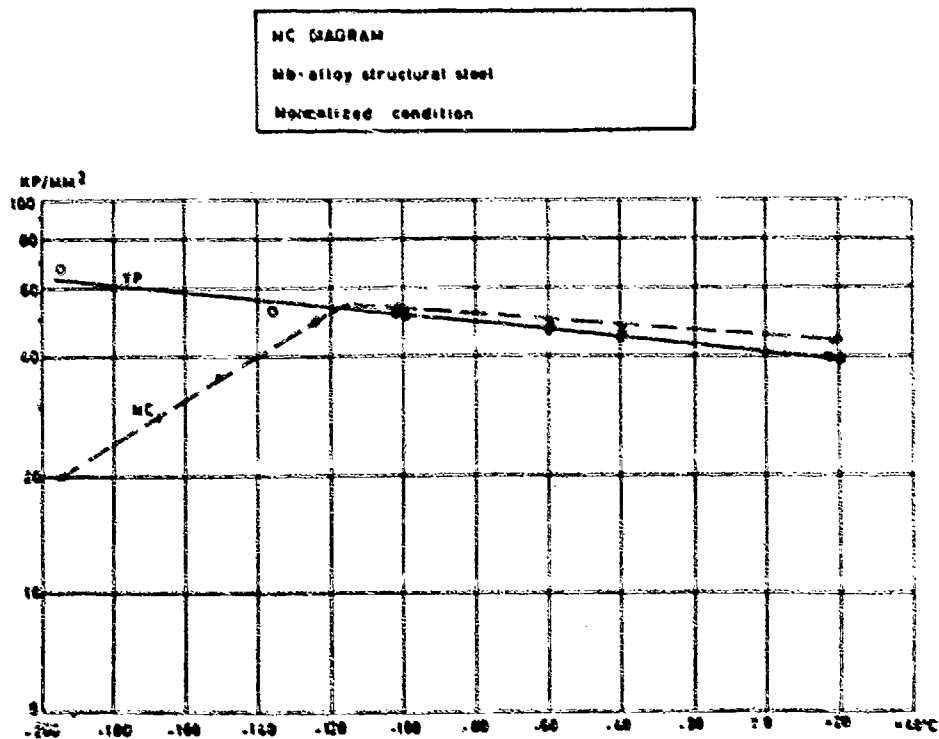


FIG. 22. NC-DIAGRAM FOR 20 mm STEEL C IN TABLE 1 IN THE NORMALIZED CONDITION. THE TRANSITION TEMPERATURE  $T_c$  IS APPROXIMATELY  $-115^{\circ}\text{C}$ .



ter is defined as the lower change of the impact curve at an impact value of about  $1.5 \text{ kgm/cm}^2$  (9 ft-lbs).

Even if it sounds peculiar a possible explanation could of course be that the columbium-alloy steel for some reason does not react as other steel types on NC-testing. In other words, the NC-testing method should perhaps not on the whole have been developed and used. I can already hear how many of my colleagues will heartily agree with this opinion. Still there is another and more reasonable explanation.

It is necessary to bear in mind that even if the NC-testing has been developed as a brittle fracture testing method it is primarily a weldability testing method with regard to brittle fracture.

The NC-test bar is evidently simulating a welded joint in a steel and the idea behind this testing principle is that the steel surrounding the welds is being attacked by sharp notches in the form of natural cracks in the welds during the testing.<sup>8</sup> Should it be so that the steel has not undergone any important change in properties around the welds the testing results will of course give a picture of the brittle fracture tendency of the more or less unaffected steel. Were it so on the other hand that a certain steel under the influence of the heat input from the welding is strongly affected that the occurred changes will have an importance concerning the brittle fracture tendency, in one direction or another, then this would be disclosed by the NC-test bar.

I have studied thousands of results on NC-testing and I must confess that generally there is practically no influence of such an importance such that one can talk about an obvious difference between the behavior of the unwelded and the welded steel. This might sound surprising but it is still a fact.

I must also admit that I got very astonished, indeed, when I compared the NC-testing results for the columbium steels with the corresponding Charpy V-notch curves. However, without any doubt it was impossible to initiate a true brittle fracture below the nominal yield strength at temperatures above  $-100^\circ\text{C}$ . Still the NC-diagrams obtained appeared quite normal with a falling NC-curve (along a straight line in the logarithmic stress scale) below the transition temperature mentioned. Therefore some good

reason for this behavior of the steel must exist.

From various other investigations we were performing at the same time two probable explanations, rather closely connected with each other, appeared.

On NC-testing in general it is well-known that even if a zone rather close to the weld will become normalized by the heat influence from the welding, this zone is too narrow to have any pronounced influence on the safety against brittle fracture of a NC-test bar. However, would this normalized zone have been a little wider, an increased resistance against brittle fracture would immediately be observed. The results would probably have become something similar to what has been observed on NC-testing of columbium-alloy steels.

However, a normalized zone alone will never be a complete explanation of the behavior of the columbium steels. It has rather been shown that on welding such steels a comparatively wide zone with increased brittle fracture resistance will appear surrounding the weld.

The background is the following investigation:

On heating a columbium steel up to  $800^\circ\text{C}$  or higher, a more or less pronounced normalization effect will occur in a quite normal way. Around  $900^\circ\text{C}$  it seems the normalizing effect has reached an optimum with regard to the ductility of the steel.

Further, however, we have found that on simultaneous plastic deformation of the steel at elevated temperatures the steel can be still more improved according to, for instance, impact properties, and this is not only limited to a rather narrow temperature range.

Our investigations, involving a certain plastic deformation at temperatures from  $750^\circ\text{C}$  up to  $1000^\circ\text{C}$ , have shown that a marked improvement of the impact strength and a still more pronounced decrease of transition temperature will have occurred after deformation and heating at  $800^\circ\text{C}$ .

The optimum of the improvements gained seems to occur in the temperature range of  $850-900^\circ\text{C}$ .

It is well-known, of course, that around a weld a plastic deformation, which certainly cannot be neglected, will take place. Conse-

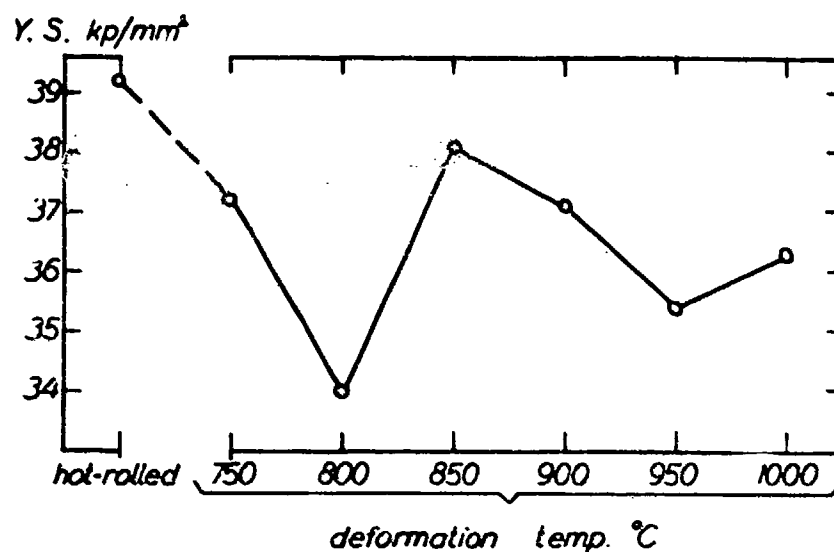


FIG. 23. DIAGRAM SHOWING VARIATION OF YIELD STRENGTH OF 20 mm STEEL C IN TABLE I IN THE AS-ROLLED CONDITION BUT AFTER PLASTIC DEFORMATION 5-10% AT VARIOUS TEMPERATURES, FOLLOWED BY AIR-COOLING.

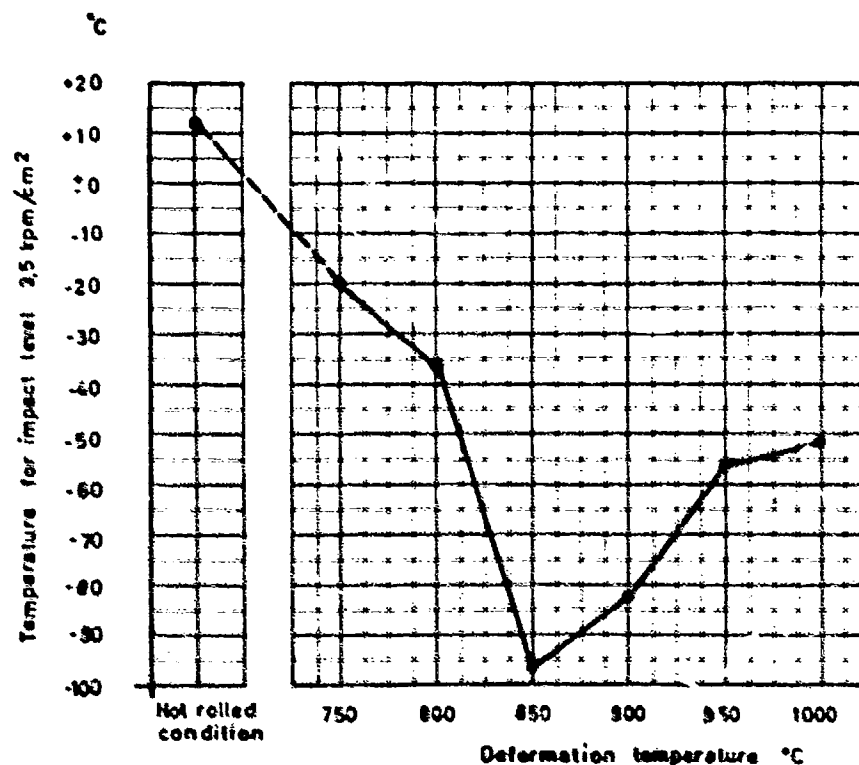


FIG. 24. DIAGRAM FOR THE SAME STEEL AS IN FIG. 23 SHOWING THE CHARPY V-NOTCH TRANSITION TEMPERATURE DEFINED AS 20 FT-LBS FOR THE SAME CONDITIONS AS INDICATED IN FIG. 23.

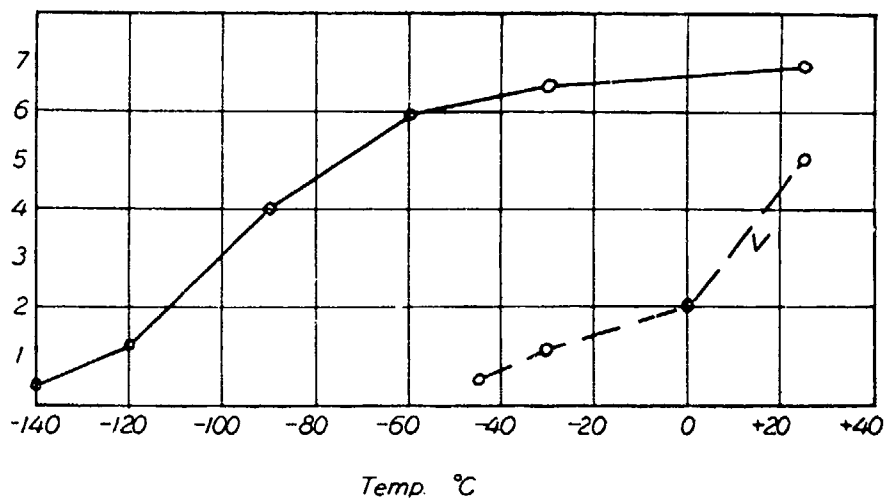
K<sub>CV</sub> kpm/cm<sup>2</sup>

FIG. 25.

CHARPY V-NOTCH TRANSITION CURVES FOR THE SAME STEEL AS IN FIG. 23 AND 24 IN THE AS-ROLLED CONDITION (RIGHT) AND AFTER 10% PLASTIC DEFORMATION AT 850°C FOLLOWED BY AIR-COOLING. THE DECREASE OF THE LOWER TRANSITION TEMPERATURE HAS A GOOD CORRESPONDENCE WITH THE LOWER TRANSITION TEMPERATURE ON NC-TESTING. A SIMILAR PLASTIC DEFORMATION IN CONNECTION WITH THE WELDING OF THE EDGES OF THE NC TEST BARS CAN BE EXPECTED IN THE HEAT-AFFECTED ZONES.

quently it can be expected that not only a narrow normalized zone with good impact properties but a rather wide zone with still better impact properties will be surrounding a weld in a columbium-alloy steel.

The results of this investigation are exemplified by Fig. 23-25. It will be seen that transition temperatures of the same size of order as observed on NC-testing have been found.

Further, and closely connected to the above-mentioned experiments at higher temperatures, we have also found that a rather strong improvement of columbium steels can be realized after tempering the steel at 500-600°C. This will still broaden the "safe" zone around a weld in this steel.

The discovery of this particular behavior of columbium steels offers indeed an improvement regarding the weldability of this material type. It also underlines that brittle-fracture testing with regard to welding technology is not always attained if the effect of welding is not included in the testing method. In other words, the NC-testing method might be able to offer at least some sort of useful information about the function stability of welds in steel with respect to the brittle fracture tendency,

particularly since the testing method is based upon the most severe defects to be found regarding brittle fracture initiation, in welds, namely transversal weld cracking, which is directly attacking the transformation zone of the base metal.

There is another advantage connected with columbium steels in welding technology. During recent years it has been shown by means of investigations in various countries that many steels used for welded structure will undergo a certain embrittlement during stress-relieving treatments at temperatures around 600°C. So far as I know this phenomenon was in Germany related to low-carbon high-manganese steels. We have been able to confirm these results but we are not willing to underline the danger of such an embrittlement so strongly as have some German investigators done, and we do not correlate it with high-manganese contents.

Apart from this difference in opinion we have been able to show that even if a carbon-manganese steel or a plain carbon steel may suffer from such an embrittlement tendency, the phenomenon can be very strongly pronounced in certain low-alloy steels. This is due to changes in the carbon distribution and partic-

ularly to carbon concentrations (not carbide concentrations) along subgrain boundaries.<sup>2</sup>

The columbium steels, which we have investigated, have of course been subject to corresponding investigations. It is quite evident that they have no tendency to temper embrittlement at stress-relieving temperatures. Further, on comparing a series of plain carbon steels, C-Mn-steels and corresponding aluminum- and columbium-treated steels<sup>13</sup> it appears that the contents of soluble aluminum may be the main cause of the embrittlement on stress-relieving rather than high-manganese contents, and that a columbium addition will inhibit this effect almost completely. For example, the Charpy V-notch 20 ft-lbs level was raised after annealing 24 hrs at 650-700°C in the order of 30-40°C, depending on soluble aluminum contents, but not at all for columbium steels. As was shortly mentioned previously, there is rather an improvement in ductility after annealing columbium steels at suitable temperatures in the range of a normal stress-relieving treatment. This is still an advantage from the welding technology point of view.

To summarize, columbium steels, in which columbium has been added as a micro-alloying element to a plain carbon steel or a carbon-manganese steel, have properties which many times are better from the welding point of view than what can be expected after testing the unwelded steel. This calls for an intensified investigation program in order to confirm such a rather unusual behavior of a structural steel.

Still it sounds surprising that a certain steel type will not reach its best properties until it has been subjected to a series of treatments in connection with welding, which are in most cases supposed to impair the steel or, under good conditions, keep it practically unchanged. It is true that a sufficiently wide experience of columbium steel in welding fabrication may still be failing, but our experience until now seems to have confirmed all what has been said above in any respect.

#### APPLICATION OF COLUMBIUM STEELS TO WELDING FABRICATION

I shall not go deeply into what can be said about columbium-steel application to welding fabrication. These steels can without any doubt be recommended to be put into welding

technology.

There are still some investigations to be performed before we can feel quite familiar with this unusual type of material, but still columbium steels have already been used in welding fabrication of various kinds--the most well-known probably being pipelines for oil and gas. Hence it would also be rather natural if somebody would like to apply the columbium steels as structural steels in a general meaning as to bridges, house-building, etc. I also feel quite certain that it will not be long until they will appear in the pressure vessel fabrication.

So far as we can see now, two different grades according to yield strength may rather easily be produced, the minimum upper yield strength of these grades being about 37 and 42 kg/mm<sup>2</sup>. Without normalizing the impact properties are today sufficient up to 1/2 inch plate thickness and by using normalizing or other heat treatments, for the present necessary above 1/2 inch plate thickness, impact properties of these grades can be guaranteed as for instance 20 ft-lbs at -30°C/-40°C. The various properties described above have shown that from the weldability point of view the columbium steels may be regarded as having at least the same weldability as the corresponding plain carbon steels and carbon-manganese steels, in some respects, however, being superior to these material types in sections, equivalent in strength.

Personally I am quite convinced that instead of increasing the yield strength of carbon-manganese steels to the bitter end, it would be wiser to take the step over to the micro-alloy steels, for instance the columbium steels, by which a yield strength, which can hardly be reached in carbon-manganese steels, can easily be obtained.

There is also a tendency within the shipbuilding industry to start using steels with higher strength than that of the present ship steels. Actually it has already been seriously discussed to present new regulations, based on new specifications for high-strength shipbuilding steels.

I have a feeling, unfortunately, that the trend today is to be rather careful on suggesting an increased yield strength. The step to be expected in this direction will probably not

lead to what is normally called high-strength steels but rather to something which can be regarded as a quite normal yield strength level for carbon-manganese steels with a minimum tensile strength of 50 kg/mm<sup>2</sup>. In other words, having for years used more or less ordinary carbon steels for shipbuilding, the shipbuilding engineers are now prepared to take the first step into an already well-known steel group, the carbon-manganese steels, which have been used for years in connection with welded structures.

This statement has by no means any ironical meaning. On the contrary I think the shipbuilders have been very wise by avoiding to introduce the carbon-manganese steels with a rather high strength in shipbuilding industry. Anybody, who has an experience of welding in shipbuilding, must be aware that the conditions under which welding at the shipyards many times has to take place, does not permit the careful handling that the carbon-manganese steels need because of their limited weldability at plate thicknesses above 25-30 mm. The susceptibility to hardening phenomena in the heat-affected zones in connection with welding is indeed more than well-known. Even if it is generally said that welding without preheating may be performed up to plate thicknesses around 25 mm, too many cases of hydrogen embrittlement in martensitic zones along welds in these steel types have occurred already at plate thicknesses far below 25 mm. Moreover, this has been the case under welding conditions, which have been far easier to control than any welding operation in shipbuilding, for instance in connection with pressure vessel fabrication.

One can refer to quite a few such occasions in Europe and I feel rather worried to have them repeated in shipbuilding, and probably more frequently.

Nevertheless the carbon-manganese steels are very much used in various branches of welding and I certainly do not want to exclude them as steels suitable for welded structures. I simply want to point out that when we have been forced to increase the strength of a certain steel type to such a degree that the weldability calls for special attention, I would not recommend such steels to be used under conditions where this special attention cannot always be paid. This is indeed not said in order to underestimate the welding engineering

in shipbuilding - it is simply a way of realizing that some parts of welding fabrication can permit strictly controlled welding conditions and others cannot. We have to regard this as a fact that must not be overlooked, and remember that the weldability of a steel always must be considered in relation to the welding conditions which can be applied.

Apparently, however, there is a need to adopt steels in shipbuilding which are higher in strength but still weldable in the sense that is to be connected with ship welding. If there are not steels of this type suitable for the purpose, let us wait and see and try to develop them rather than to apply well-known types of the desired strength, which are not only well-known from previous successful applications under certain conditions but unfortunately also well-known from unsuccessful applications under the conditions to be considered.

It is easy to say that one should wait and it is also easy to say that one should develop a new steel type--the requirements of which must be rather high. I must admit that I am not quite certain if I would have said what follows five or six years ago, before the first results of the micro-alloying technique had appeared.

Nevertheless, today I would definitely recommend the shipbuilding industry to be more careful regarding the weldability of new high-strength steels rather than the increase in yield strength. In other words, as said previously, it is my opinion that it should be wiser to specify a yield strength of the new steels that will not permit the use of carbon-manganese steels but rather requires some type of a micro-alloy steel, whatever the micro-alloying element may be.

This must not be regarded as a recommendation to start using columbium micro-alloy steels or any other type too soon. It should rather be recommended that such steel types ought to be intensely investigated with regard to any property that has an importance in connection with welding in shipbuilding. This may take a little time, but it appears to me more realistic to be interested in new and promising steel types than to believe in steels, which have already proved to be less suitable for unfavorable welding conditions.

It also occurs to me that what the shipbuilders are looking for with respect to in-

creased strength, safety against various treatments and reasonable costs will probably be met by some micro-alloy steels with a minimum upper yield strength of about 37-45 kg/mm<sup>2</sup>. Such steels will obviously meet the strength requirements and are in the same time from the weldability point of view placed in the preferred part of the steel group to which they belong because of the comparatively low-strength level within this.

#### COLUMBIUM AS PART OF COMPLEX STEEL ALLOYS

It is not my intention to go deeply into a lot of complex steel compositions, in which columbium is one of the important elements. I would like to mention, however, that our interest in columbium steels has not been limited to micro-alloying structural steels with columbium as the only micro-addition.

It has been mentioned above that columbium steel from the yield strength point of view may cover a range of 37-45 kg/mm<sup>2</sup> and it has also been said that increasing columbium contents is not the main mean to raise the yield strength within this group of steels. It is rather so that from many points of view it is preferable to keep the columbium contents rather constant, at about 0.02-0.03%, and change the carbon contents and/or manganese contents in order to reach various strength properties. This will limit the weldability of the steel group in the way that at an upper yield strength level of 45-50 kg/mm<sup>2</sup>, the carbon and manganese contents have had to be increased to such a degree that a further increase will drastically limit the weldability. In order to reach still higher yield strength levels and, of course, in the same time keep other properties as much unchanged as possible, it has proved necessary to develop more complex micro-alloy steels.

In the first place we have concentrated on two further micro-alloying elements to be used in connection with columbium, aluminum and vanadium. See also Appendix C.

A combination of columbium and aluminum, as far as our experience is concerned, can be summarized very shortly. It is doubtful and peculiar. Until now we have not been able to disclose why our results have become what they have become. On adding various amounts of aluminum of the same size of order as for a normal fine-grain aluminum treatment to a steel

to which columbium has also been added, one will find that the well-known phenomenon of formation of aluminum-containing sulphides is very much pronounced. For some reason the aluminum addition will completely change the sulphide inclusions as shown in Fig. 26. Such a sulphide distribution with long tiny aluminum-containing sulphides has a most detrimental effect on the bending properties of the steel as well as on the elongation. The steel will show what has been called a pronounced "short breaking behavior". We have investigations still running on this type of micro-alloy steel and in some cases we have been able to overcome this detrimental aluminum effect but we cannot always reproduce the heats which have come out successfully, and, more important, no particular advantages have been found.

We have been more successful by using a combination of columbium and vanadium as micro-alloying elements. It has proved that in order to increase the yield strength level above what can be reached by columbium only within reasonable weldability limits, a further addition of vanadium in the same size of order as the columbium addition will extend the yield strength range with another 5-8 kg/mm<sup>2</sup>, while the weldability seems to remain mainly unchanged.

Besides an increased yield strength, a vanadium addition will also cause a precipitation hardening on tempering the steel within the temperature range of 500-600°C. This is an effect corresponding to that of columbium which a further addition of vanadium will increase. Therefore there may be some advantages connected with columbium-vanadium steels which cannot be reached by columbium steels only (Fig. 27).

Another effect of vanadium is, for instance, that after normalizing a columbium-vanadium steel will not show the same strong decrease in yield strength as if columbium was the only micro-alloying element. Consequently in practical production work columbium-vanadium steels may many times be preferred simply because heats too high in yield strength can be normalized in order to fulfil the maximum yield strength specified, while charges too low in yield strength can be tempered at a suitable temperature in order to fulfil the minimum yield strength specified. In both cases the impact properties of columbium-vanadium steel are improved and, actually, this steel type should



FIG. 26. LONG TINY SULPHIDE INCLUSIONS IN A MICRO-ALLOY COLUMBIUM STEEL AS A RESULT OF A CRITICAL ALUMINUM ADDITION. 200 x

always need some sort of heat treatment in order to get a desired safety against brittle fracture. However, in this paper I am not supposed to outline the influence of columbium in complex micro-alloy steels but rather this element as a steelmaking variable itself. Therefore I shall not discuss low-alloy steels with columbium additions either, but only mention that we have a certain interest in manganese-molybdenum steels and molybdenum-copper steels with small columbium additions, the most pronounced influence of which is an improvement of the impact properties and a stabilizing effect on the strength properties on the whole after treating at normal stress-relieving temperatures.

There are various philosophies to be applied to metallurgical aspects on welding technology. Nobody would today be able to say which of them is to be regarded as the best one. Personally I feel that no one should be called more correct than anyone else provided we are dealing with those on which the enormous development of the welding technique has been based - and we must remember that there are quite a few of them.

In technical considerations, however, there is one matter of particular importance - consequence. This means consequence in making, rolling, treating, controlling and using materials, regarding steels as individuals as to their behavior in a structure - in other words, consequence in thinking.

A way of thinking, implying to collect the very best of the best of methods, materials and calculations will have no sense if the consequence is failing. We should never let our ambition to do the best prevent us to do something good.

I shall finish by quoting a colleague and very close friend of mine, who has recently retired, Mr. T. W. Bushell, former Principal Surveyor for Metals of Lloyd's Register of Shipping. Some years ago, during a discussion after a lecture concerning the brittle fracture problem, he summarized his thoughts by saying "The explanation of the fracturing behavior of a ship steel should primarily not be a matter of too much personal research prestige but rather a problem of measuring the safety for the men who sail our ships."

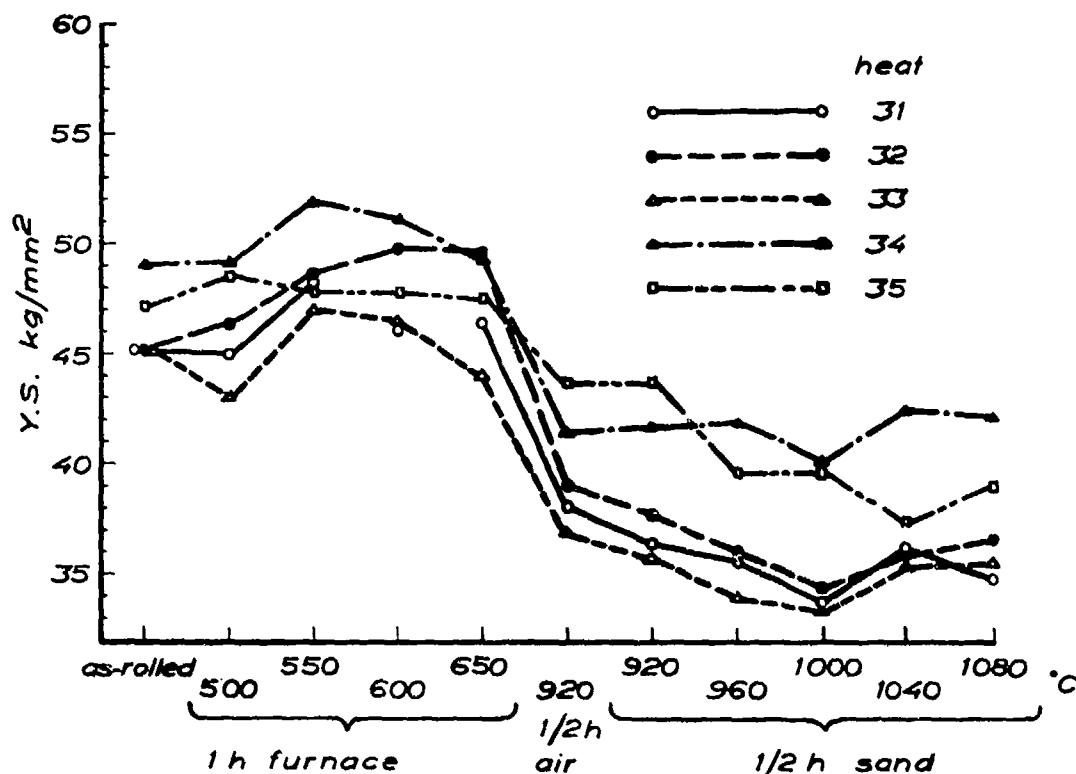


FIG. 27. YIELD STRENGTH VERSUS VARIOUS HEAT TREATMENTS AFTER ROLLING OF THE FOLLOWING HEATS:

No.	C	Si	Mn	P	S	V	Cb
31	.16	.09	1.23	.06	.052	-	.05
32	.15	.08	1.28	.06	.052	.03	.05
33	.19	.09	.85	.059	.051	.05	.04
34	.16	.07	1.26	.059	.053	.10	.05
35	.21	.11	1.39	.032	.038	-	.056

THE DIAGRAM INDICATES THE YIELD STRENGTH IN THE AS-ROLLED CONDITION (TO THE EXTREME LEFT) AND AN INCREASED YIELD STRENGTH AFTER TEMPERING BETWEEN 500 AND 650°C DURING 1 HOUR, FOLLOWED BY COOLING IN THE FURNACE. THE DIAGRAM FURTHER SHOWS THE CHANGE IN YIELD STRENGTH AFTER NORMALIZING AT 920°C DURING 1/2 HR., FOLLOWED BY AIR-COOLING AND A FURTHER CHANGE IN YIELD STRENGTH AFTER HEAT TREATING DURING 1/2 HR., FOLLOWED BY COOLING IN SAND FROM TEMPERATURES WITHIN THE RANGE 920-1080°C.

This is a necessary and important statement based on human experience at its best.

#### ACKNOWLEDGMENT

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#### APPENDIX A

##### THE NWH-TESTING METHOD

In 1952 a Weld-Hardening test was developed by the Author in co-operation with Mr. G. Bini.<sup>14</sup> The following description of the method refers almost in detail to a later publication in English.<sup>1</sup> At the beginning the method was known as "HF-testing" but has later become known as the NWH-testing (Norén Weld-Hardening test):

A hardenability test was developed at the ESAB laboratories in Sweden, for use in cases in which both the heating and cooling processes are rapid. It has proved to be highly reproducible and to correlate well with actual welding experience. The following describes in detail the principles underlying the test, and the experimental technique.

The rapidity of the welding process does not allow complete diffusion and transformation to austenite to occur in the heat affected

zone. For this reason the Jominy test is not ideal, as far as welding is concerned and in fact the microstructure and hardness found in the heat-affected zone of a weld, frequently differ from those in the corresponding area of the Jominy test bar.

The rate at which a Jominy bar is heated is slow compared with welding conditions, and this leads to complete transformation and homogenization of the austenite. In welding, on the other hand, the heating period is only 3 to 15 seconds; diffusion is incomplete and certain alloy carbides, for example, are not taken fully into solution.

Again a Jominy test bar for weldability testing is austenitized at temperatures about 1100°C, whereas in the heat-affected zone the maximum temperature reached is the melting point of the material--a fact of primary importance, as regards the austenitization process.

Thus in the heat-affected zone austenitization temperatures are found over the entire range from  $A_1$  to the melting point and at its outer extremities the steel will transform only partially to austenite. The Jominy test can thus only refer to one very small part of the heat-affected zone, whereas all parts are of equal interest and especially the partially transformed areas.

Moreover, since a Jominy test bar is held at austenitizing temperature for an appreciable period, the prior structure of the material has little or no influence on the result. In welding, on the other hand the prior structure is of vital importance for the behavior of the heat-affected zone. A hardenability test designed for the study of welding problems must reproduce these conditions of rapid heating and short-time soaking at all temperatures up to the melting point.

A further point of perhaps secondary importance is that the heat flow during cooling in the heat-affected zone is always from the more superheated parts to the less (with a minor side-loss to atmosphere from the surface). This is not truly reflected in the Jominy test.

The size of the original Jominy test bar makes it unsuitable for testing material under 1 inch thick. Certain modifications have been suggested<sup>1</sup>, but it is still difficult to adapt the test to material below 1/2 inch. A harden-

ability test for welding applications should utilize smaller specimens, preferably not requiring material more than 1/4 inch thick.

Finally, the Jominy test is not very suitable for testing low-carbon steels or the shallow-hardening types of high-carbon steel.

The Jominy test has been a valuable aid in welding research. However, there is a need for an improved hardenability test, designed specifically for weldability studies, particularly with the increasing use of low-carbon weldable alloy steels. It must be remembered, that the Jominy test has been of greatest value in connection with the repair welding of fairly deep hardening high-carbon steels.

It would be unfortunate if the calculations relating welding conditions to the Jominy test were to be rendered obsolete. It is reasonable therefore in developing any new test to correlate it as far as possible with the Jominy test so as to make full use of the valuable and very extensive data which are already available.

All these requirements can be met by using a single small test specimen heated by high frequency, which is an inexpensive but accurately controllable heating method. It is, however, apparently necessary to test the specimen not once but repeatedly under different cooling conditions. This type of test was under investigation for some years in the ESAB laboratories, with the aim of setting the optimum conditions. The heat source used was a 1 KW high-frequency generator.

A suitable test bar size is 5 x 5 x 150 mm. The heating period finally selected was 6 seconds, the generator being adjusted to bring the high temperature end of the specimen just up to melting point. The procedure finally arrived at is as follows:

As is shown in Fig. 28 the test piece is placed vertically with one end in a bath of water maintained at +15°C (+5°C) leaving a "free length" of variable length  $F$  projecting above the water level. The top of the test piece is level with, and central in the heater coil. After a 6-seconds heating period, when the top end should have just reached melting point, the generator is automatically cut out. The heating cycle sets up a temperature gradient down the bar, from melting point down to the water bath temperature, and in the cooling

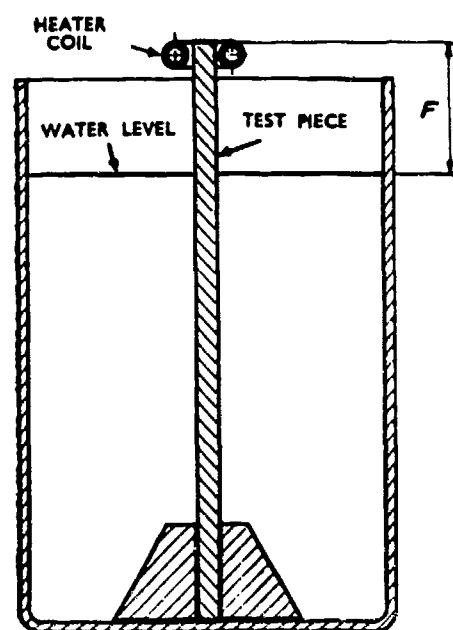


FIG. 28. TEST BAR ARRANGEMENT FOR NWH-TESTING. THE CROSS SECTION OF THE TEST BAR IS 5 x 5 mm. APP. A.

cycle all the heat flows down to be absorbed in the bath. Thus the initial temperature gradient and the cooling rate depends on the value of  $F$ , the free length.

A series of tests is carried out, reducing the free length for each successive test; the values selected are quite arbitrary, the Author having used series such as:

- 5 - 7 - 8 mm
- 5 - 10 - 12 mm
- 5 - 15 - 25 - 50 mm, and so on.

When the first test has been made, the free length of the bar is cut off at or just below the water level, where temper coloring can be seen. The remainder of the test bar is set up at the new free length, fresh water being added to adjust the water level and so the tests are continued.

The cut lengths are mounted in bakelite and a longitudinal flat is ground as in the Jominy test to a depth of 0.5 to 1 mm. The grinding must be carefully carried out under a copious stream of water to prevent any tempering of the martensite. The test flat is finally polished, and Vickers hardness readings are made with a 10 kg load at intervals of 0.5 to 1 mm along the center line of the flat. Where necessary the

results can be amplified by micro-hardness testing in the areas where microscopic examination suggests local hardness peaks; these results will clarify the macro-hardness curves.

- With experience in the method it is possible to predict within limits the probable behavior of a material and restrict the test to those  $F$  values which might be considered critical.

Because of the use of high-frequency heating, this test has at the beginning become known as the HF test and later as the Norel Weld-Hardening test.<sup>1</sup>

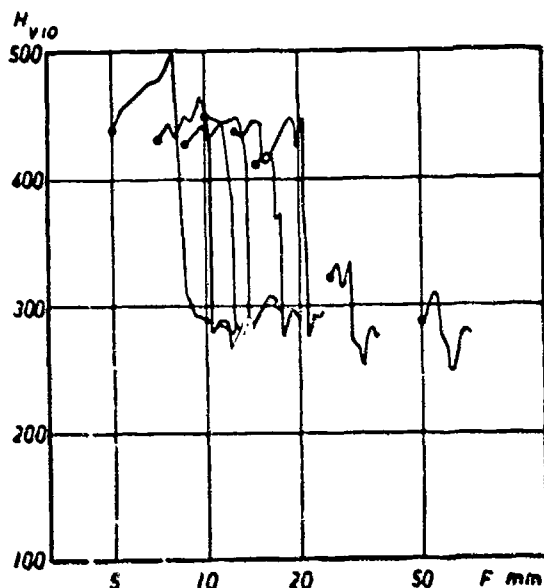


FIG. 29. SECONDARY DIAGRAMS ACCORDING TO THE NWH-TESTING METHOD FOR A LOW-ALLOY HIGH-STRENGTH STRUCTURAL STEEL. (VICKERS HARDNESS ON 10 kg LOAD VERSUS  $F$ -DISTANCE. CORRESPONDING TO  $2.5 \times$  JOMINY-DISTANCE - 1.) APP. A.

A complete hardenability diagram as given by the NWH test consists of a set of hardness curves for the various  $F$  values, as shown in Fig. 29. The diagrams are similar to Jominy curves with hardness on the vertical scale and cooling rate horizontally, in terms of the free length  $F$ , which corresponds in principle to the distance from the quenched end in the Jominy test. The hardness curves for the different  $F$  values are known as secondary diagrams, and show the hardness variations occurring in the heat-affected zone at a given constant average

cooling rate. The fusion line hardness value in the base metal adjacent to the molten zone is plotted at a point corresponding to the value of  $F$  as shown by the black dots in Fig. 29. The remaining hardness readings for this particular test are plotted to the right of the fusion line hardness, at distances corresponding to the distances along the test bar. When all the secondary diagrams have been plotted (Fig. 29 contains more than are usually required) the various fusion line measurements (black dots in Fig. 29) are joined by another curve which is known as the primary diagram and corresponds to the Jominy curve (Fig. 30).

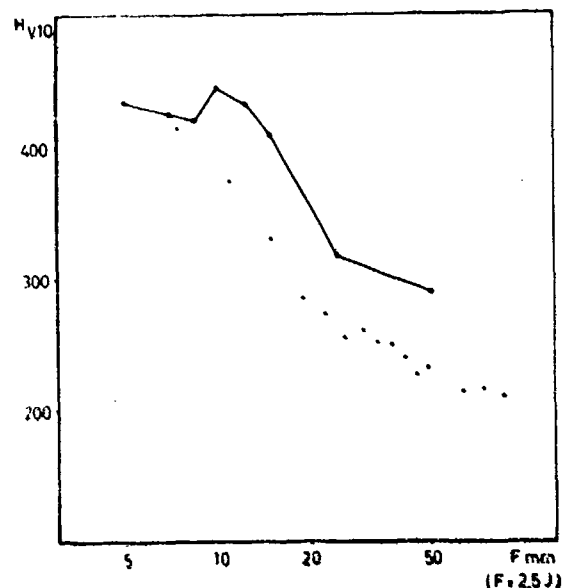


FIG. 30. PRIMARY DIAGRAM ACCORDING TO THE NWH-TESTING METHOD FOR THE SAME STEEL AS IN FIG. 29. THE CORRESPONDING JOMINY CURVE OF THE STEEL AFTER AUSTENITIZING AT  $1100^{\circ}\text{C}$  IS REPRESENTED BY THE DOTS. (VICKERS HARDNESS ON 10 kg LOAD VS.  $F$ -DISTANCE. CORRESPONDING TO  $2.5 \times$  JOMINY-DISTANCE - 1.) APP. A.

It has been possible to correlate the  $F$  values empirically with Jominy distance. Under the conditions given (particularly of test bar size) the correlation factor is:

$$F = 2.5 J.$$

It is obviously convenient to select  $F$  values which are multiples of 2.5, so that all the calculations made on the Jominy test are directly applicable to the NWH test.

With the NWH test the heating and cooling processes in welding are more nearly reproduced. A test which shows the variation of maximum hardness with cooling rate will also indicate the hardness variations with a given heat-affected zone cooled at a given overall rate. Furthermore the hardness readings can be correlated with microscopic examination of the microstructures obtained under various heating and cooling cycles.

The test piece is small enough to be applicable to most steel products, and the test is accurate, highly reproducible and, in spite of the preparation required for hardness testing, rapid.

The test conditions as described are the outcome of extensive experimental work. Good agreement is found between the NWH primary diagram and the Jominy curve for steels which transform completely to austenite, at least in the hottest zone, during the 0.5-seconds heating period. This is true in particular for carbon steels of about eutectoid composition with uniformly dispersed pearlite. Alloy steels containing stable carbides exhibit differences between NWH and Jominy curves which are explained by the incomplete solution and diffusion of these carbides in the short period available. As a result the hardness maxima vary in the most rapidly cooled test pieces (since the austenite composition is different), and frequently the steels appear to be deeper hardening, because the austenitizing temperature is higher.

A special advantage of the NWH test is its ability to distinguish between steels of similar conventional properties, e.g. of equal ultimate tensile strength. Different effects are produced not only by small composition differences but also by differences in prior structure. Laminar inhomogeneities in rolled plates, for example pronounced ferrite banding, lead to local hard spots in partially transformed zones.

## APPENDIX B

### THE NC-TESTING METHOD

The following refers to parts of a previous paper:<sup>8</sup>

#### Testing Principles and Testing Method

It is assumed that the reader is familiar with the principles of the two types of plastic deformation of steels and other metallic materials. It is beyond the scope of this paper to detail the two possible mechanisms of deformation but it may be stated that the translation (gliding) mechanism can cause a considerable deformation, while twinning preferably takes place under complicated stress conditions and/or at low temperatures; from a practical point of view the latter is insignificant.

In the following the term "brittle fracture" refers to a cleavage fracture without appreciable preliminary plastic deformation of the individual crystals. For a tensile test this means that a brittle fracture has occurred at a nominal stress below or only slightly above the conventional yield point. In the first case plastic deformation has taken place only at the fractured section and is hardly measurable. In the second case a small amount of plastic deformation has occurred in a large material volume of metal and the fractured surface appears "crystalline", i.e. the cleavage planes of the individual crystals are clearly seen in the surface.

To obtain fracture as near as possible without deformation at temperatures where uniaxial stress gives rise to deformation, the occurrence of local and complicated triaxial stresses in the material is necessary; for example, this occurs in the presence of a sharp notch: at the root of the notch even low nominal stresses can cause local plastic deformation, while the remainder of the material that is unaffected by the multiaxial stress will remain largely undeformed. The severest notch effect is caused by a crack in the material.

When loaded the material will deform at the crack front. At low nominal stresses this deformation can be so important that the degree of triaxiality in the stressed condition is sufficiently reduced to eliminate the risk of fracture. In consequence the notch radius at the crack front will increase as a result of the deformation and will raise the level of triaxial stress, which can cause crack propagation. With a continuously increasing load, the increase of crack radius is relatively quicker and a new crack is necessary for a critical stress condition to develop at a higher load. If a sharp crack front is always present in the material, however, the triaxial stress condition at a certain critical nominal stress may be such that the material cannot resist a further increase in stress even though locally deformed. Cleavage of the crystals at the crack front then takes place and the fracture propagates more or less rapidly.

By means of a brittle alloy welded on the edges of a flat tensile test bar, the conditions mentioned can be realized while cracking is continually occurring in the weld metal, as the bar is exposed to an increasing tensile stress. At a certain critical stress, depending on the testing temperature, a more or less brittle fracture is obtained from one of the sharp crack fronts present. This critical stress, which is defined as the highest nominal tensile stress a steel can maintain in the presence of an undeformed crack without the initiation of a propagating fracture, is the so-called nominal cleavage strength. It appears to the author that this is a characteristic property of a steel and might be useful in strength calculations. Considering the problems associated with the brittleness phenomenon of a steel, the author has stated the following seven points as a basis for NC-testing:

1. The conventional yield point of a steel rises continuously and the plasticity at translation falls continuously with decreasing temperature.
2. Above a certain nominal loading the initiation of a fracture at a sharp notch, such as a crack, can only be prevented by plastic deformation of the crack front.\*
3. The less the plastic deformation, the lower the nominal load required to give a stress condition at the crack front that will cause a cleavage fracture.
4. In consequence the maximum nominal load that a steel can sustain without cleavage fracture at a crack front, i.e. the "nominal cleavage strength," decreases with falling temperature.
5. Further, as the cleavage strength depends on the plasticity, it must decrease continuously with falling temperature.
6. Curves showing the rise of yield strength and the decrease of cleavage strength with falling temperature must intersect at some temperature.
7. The intersection point of the curves can be taken as the transition temperature of the steel above which a fracture can start only after the occurrence of plastic deformation in a large volume of material but below which the fracture start and progress require negligible local plastic deformation in the fractured section (i.e. brittle fracture).

In a similar way to the yield point, the nominal cleavage strength is a characteristic of every steel for a certain strain rate, depth of notch and sharpness of notch. As previous-

\* By "nominal loading" is meant the stress calculated on the total cross-sectional area of the test piece (including welds) until the welding fractures. As soon as a crack starts in the welding all the applied load is transferred to the total cross-sectional area of the welded test piece less the area of one of the welding layers. The investigations have shown that the chance of fractures occurring in both welding layers at the same time and opposite each other is very small. It can be added that the correction due to the layers of welding has little effect compared with the normal divergences in, for example, the yield point of normal test pieces taken in different positions in the test material. No great error will occur in practice if the calculations are made with the area of the test piece before welding. However, such an error should not be made even if it has little effect.

ly stated it can be expected that the cleavage strength, as the yield point, will vary continuously with the temperature, provided that the other conditions of testing remain constant. At the defined transition temperature, the cleavage strength coincides with the yield strength. Later it will be shown, however, that this transition temperature called  $T_c$  does not represent the one below which brittle fractures occur and above which only fibrous fractures (shearing fractures) take place. Typical crystalline fractures are also observed at temperatures slightly above  $T_c$  and obviously (this is discussed later) another and higher transition temperature called  $T_s$  exists above which nothing but shearing fractures are observed. In principle  $T_c$  and  $T_s$  correspond to the two change points of a temperature-impact-strength curve. The dimensions of the NC-test bar and the welding of its edges are shown in Fig. 31; the rate of loading used is 10 mm/min., and before testing the test pieces are cooled in a suitable medium such as solid carbon dioxide, liquid oxygen or the like. The temperature is measured by a gauge, which is

thoroughly isolated from the influence of the surrounding air, and the temperature is registered by a resistance thermometer. Before testing, the specimen, fitted in the grips, is allowed to warm up until the desired temperature is nearly reached, whereupon the load is applied.

### Testing Results and the Relationship between NC-Testing and Other Methods

A great number of static NC-tests have been carried out by the present author and others, and it has been found that the critical temperature at which the yield-strength curve and the curve of the nominal cleavage strength interest,  $T_c$ , is located without exception within the same temperature interval as the lower change point of a Charpy V-notch curve. As far as ordinary low-carbon structural steels are concerned, the transition temperature referring to NC-testing corresponds to a numerical value of the Charpy V-notch impact strength between 1 and 3.5 kgm/cm<sup>2</sup> (about 6-20 foot-pounds). The critical impact strength value, however, is not independent of the yield strength of the steel. (This will be shown later.) Fig. 32 shows the relationship between the NC curve, the yield-strength curve and a Charpy V-notch curve including the

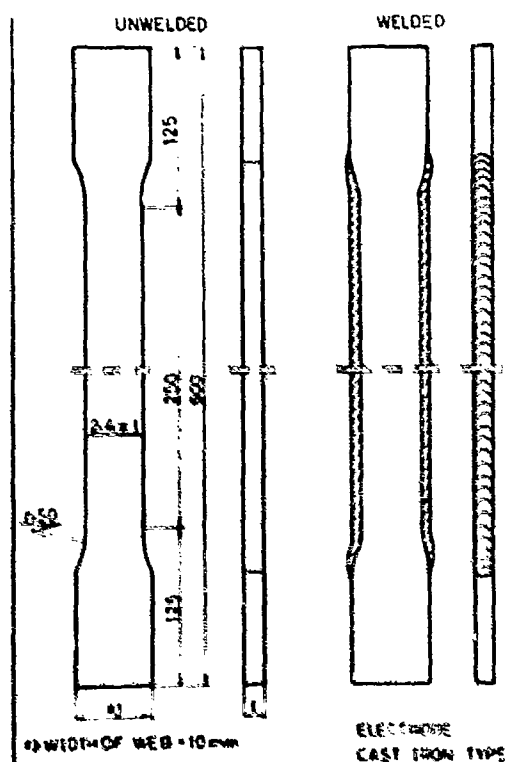


FIG. 31. TEST BAR TYPE USED FOR NC-TESTING. APP. B.

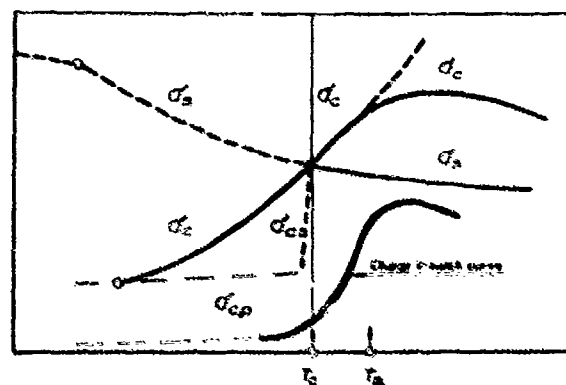


FIG. 32. DIAGRAM SHOWING IN PRINCIPLE THE RELATION BETWEEN THE YIELD STRENGTH CURVE, THE NC CURVE (THE NOMINAL CLEAVAGE STRENGTH), THE CURVE FOR PROPAGATION OF A BRITTLE FRACTURE ACCORDING TO THE NC-TESTING METHOD AND THE CHARPY V-NOTCH IMPACT CURVE. APP. B.

position of the two transition temperatures  $T_i$  and  $T_c$ . The way the impact curve as well as the NC curve fall off at higher temperatures should be noted. In both cases this means that in calculating the energy absorption and

the nominal cleavage strength respectively, one has not paid attention to the reduction in area, and the values obtained are too low. The reduction in area, on the other hand, shows that the material at the testing temperature possesses a higher degree of plasticity. Consequently, the upper change point of the NC-curve as well as that of the impact curve simply imply that at this temperature the steel has reached a temperature range within which complete shearing fractures are obtained. An attempt to find a numerical value of the upper change point of the Charpy V-notch curve, as has been found with regard to the lower change point, has failed, a fact that Matton-Sjöberg<sup>16</sup> has underlined in connection with his interpretation of impact strength curves. Nevertheless the upper change point of the Charpy V-notch curve will always be found at practically the same temperature as the upper change point of the corresponding NC-curve. No correlation exists between the numerical value of the Charpy V-notch impact strength at  $T_1$  and the level of yield strength or nominal cleavage strength at the same temperature, however. On the other hand, where the two latter properties coincide, there is no doubt that the corresponding Charpy V-notch value falls within a comparatively narrow temperature range. The part of the impact curve between the lower and the upper change point, where the steepest slope is to be found, as well as the part of the NC-curve between  $T_1$  and  $T_2$ , represent the temperature range within which sufficient plasticity of the steel occurs to permit translation of a crack and in a large volume of material; yet the material has insufficient ability for plastic deformation and for absorption of energy effectively to stop a propagating fracture.

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Both transition temperatures,  $T_1$  and  $T_2$ , are important with regard to the tendency of welded structures to brittle fracture. Simply,  $T_1$  is the lowest temperature at which a steel can still deform plastically despite the presence of cracks and above which the welding stresses are probably of no consequence. The temperature  $T_2$  is approximately the one above which a steel may be regarded as crack-arresting due to the occurrence of true shearing fracture. Discussions concerning the importance of the lower temperature  $T_1$  in connection with welding stresses have been published by Soete, Boyd, and others<sup>10-12</sup> and by means of NC testing it has been possible to put forward

experimental evidence to support the valuable contributions of these authors (see below).

Finally; before certain features of the NC test results are interpreted, the following must be underlined as one of the most important statements concerning service failures through brittle fractures of welded as well as unwelded structures. The presence of a defect in the material, representing a severe notch-effect under the loading conditions applied to the structure, is necessary for the initiation of a fracture and for the increased risk of a service failure when the temperature falls.

In the absence of a notch, the strength of a steel will increase as the temperature falls. If a notch of the same sharpness as that used in the Charpy V-notch test is cut in a tensile test-bar, e.g. Tipper-test specimen, no detrimental influence will occur in the strength of the bar as the stress for fracture as well as the yield strength continuously increases with decrease in temperature. In most cases a notch has to have about the same sharpness as a true crack to be dangerous.

#### The Influence of Temperature on NC Curves

It has been shown that a similarity between an NC curve and an impact-strength-temperature curve exists, and therefore it might be assumed that the two curve types are influenced by the same property of the steel. An attempt to express the whole curve in terms of mathematics would probably be unsuccessful because of the complexity caused by the reduction in area of the test-bar in shear fracturing. However, it might be possible to find a simpler law expressing the shape of the curves within the temperature range where test-bars experience little reduction in area.

If a Charpy V-notch curve is compared with an NC curve, as has been done in Fig. 12, one will find that the main part of the former, i.e. the part with the steepest slope, corresponds to a comparatively small part of the NC curve, namely the one just above the point of intersection between the NC curve and the yield-strength curve. This seems to show that it is better to use the NC curves for a closer study of the part that is not influenced by the reduction in area. The seven points summarizing the basis for NC testing include the fact that the slope of an NC curve depends on the possibility of plastic deformation at the front of



the notch or crack in the welded layer. This can also be expressed by saying that the cracks in the weld can initiate a fracture if the testing conditions are such that the test-bar material has a certain resistance to plastic deformation at the crack front, i.e. if the resistance to translation is high. A low nominal cleavage strength within the temperature range below the transition temperature thus corresponds to a high resistance to translation and vice versa.

The resistance to translation depends upon certain "external" factors such as stress condition, temperature, rate of loading, etc., and also upon some "internal" properties of the steel such as grain size, dislocation conditions, presence of residual stresses and the influence on the lattice of previous mechanical and thermal history (aging, previous dynamic loading, etc.). Generally the "external" factors can be controlled by means of testing conditions, but the "internal" ones are difficult if not impossible to express by means of the usually accepted physical definitions or terms. The resulting phenomenon, which includes all the factors mentioned as well as others that are not stated, may be called the resistance to translation. It is not feasible to derive an expression for this quantity in terms of its different parts, as only a few of them can be expressed with sufficient accuracy. For the present, therefore, the following discussion will only include "resistance to translation" as a general term.

It is obvious that this property is the one influencing the position of the conventional yield-strength curve, and therefore a correlation probably exists between the influence of temperature on the upper conventional yield strength and on the nominal cleavage strength. It can be shown that the yield strength is an exponential function of the temperature and it is possible to use the same type of expression for the nominal cleavage strength. The author has simplified the well-known expression

$$Y = (ve^{Q/RT})^r \quad (1)$$

given in Ref. 29, ( $v$  = loading rate,  $T$  = absolute temperature,  $R$  = the gas constant,  $Q$  = an energy of activation,  $r$  = "a small number"), by using

$$y = a^x \quad (2)$$

The two variables are expressed as follows:

$$N_s = \frac{\sigma_{s,1}}{\sigma_{s,0}} = a_s (T_0 - T_1) \quad (3)$$

where  $\sigma_{s,1}$  is taken as an expression for the upper yield point at the testing temperature,  $\sigma_{s,0}$  as the upper yield point at the original temperature, and consequently  $N_s$  the relative yield point.  $T_0$  is taken as the origin temperature and  $T_1$  as the testing temperature at which  $\sigma_{s,1}$  has been measured. Finally,  $a_s$  is a parameter used as an expression for the resistance to translation under the prescribed testing conditions.

A similar expression can be used for the nominal cleavage strength, namely

$$N_c = \frac{\sigma_{c,1}}{\sigma_{c,0}} = a_c (T_0 - T_1) \quad (4)$$

which indicates a decrease in nominal cleavage strength with falling temperature in contrast to the increase of yield strength when the temperature decreases. Here  $\sigma_{c,0}$  means the nominal cleavage strength at the original temperature  $T_0$ ,  $\sigma_{c,1}$  the nominal cleavage strength at the testing temperature  $T_1$ , and consequently  $N_c$  the relative nominal cleavage strength. As a result of the notch effect, the parameter  $a_c$  is not similar to  $a_s$  in the foregoing equation.

When  $T_1$  coincides with  $T_0$ , the relative yield strength as well as the relative nominal cleavage strength  $N_c$  have values of unity;  $T_0$  therefore equals  $T_c$  (the point of intersection between the yield-strength curve and the NC-curve, i.e., the strength of an NC test-bar at the transition temperature as defined for the method of testing). The use of Eq. (3) and (4) to obtain approximate values of the properties of the two steels with temperature variation can easily be shown experimentally.

The characteristics of steels, having the same yield strength at the same transition temperature but different magnitudes of the parameter  $a$  are represented by the curves shown in Fig. 33. Now the term  $T_0$  can be replaced by  $T_c$ ; the slopes of the curves given by Eq. (3) and (4) can be expressed as:

$$N_s = -a_s (T_c - T_1) \ln a_s \quad (5)$$

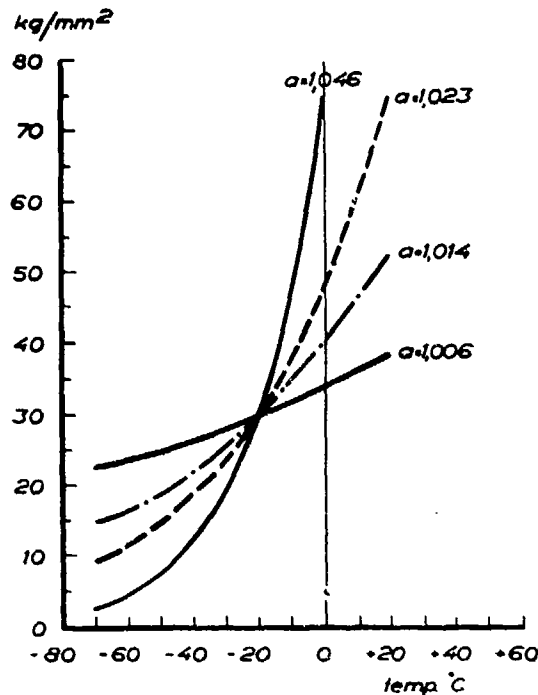


FIG. 33. SET OF NC CURVES WITH THE SAME TRANSITION TEMPERATURE  $T_c$  ( $-20^\circ\text{C}$ ) BUT VARIOUS RESISTANCE TO TRANSLATION ( $a$ ). IF THE STRESS IS PLOTTED ON A LOGARITHMIC SCALE VS. TEMPERATURE. THE CURVES WILL BE REPRESENTED BY STRAIGHT LINES ACCORDING TO THE EQUATIONS GIVEN IN THE TEXT. APP. B.

$$N_c^* = a_c (T_c - T_c) \ln a_c \quad (6)$$

Further it will be noted that the straight lines representing Eq. (3) and (4) (Eq. 7 and 8) on a logarithmic plot have the same slope as the curves at the transition temperature:

$$\ln N_1 = (T_c - T_c) \ln a_1 \quad (7)$$

$$\ln N_2 = (T_c - T_c) \ln a_2 \quad (8)$$

The validity of the above-mentioned equations (at least as approximate expressions) was checked by comparing theoretical and actual results of a large number of NC tests.

It seems reasonable to expect that the energy absorption in an impact test at temperatures below the point of intersection between the two curves of an NC diagram should be insignificant, but it is more likely that the

energy level depends on the yield strength of the material at the testing temperature. If this is so, the energy absorption in impact tests made at the transition temperature  $T_c$  (for NC testing) should decrease with the yield point of the steel. This would support the use of different minimum requirements for the critical impact strength of different steels depending on the yield strength.

One of the most important observations of the investigations described is as follows:

Evidently a steel cannot possess both a high nominal cleavage strength at low temperature and strong energy absorbing properties on fracture at higher temperature. For a steel with a flat NC curve, the nominal cleavage strength is considerable at temperatures below  $T_c$ , but the energy absorption in shearing fracture will only be important at comparatively high temperatures. On the other hand a steeply sloping NC curve indicates a low nominal cleavage strength below  $T_c$  but high energy-absorbing properties slightly above this temperature. The latter type of steel has, therefore, a lower transition temperature for shearing fractures  $T_s$ , than the former; a high resistance to translation brings  $T_c$  closer to  $T_s$  and vice versa.

#### Interpretation of NC Diagrams

The following interpretation of an NC diagram is based on the assumption that the continuous NC curve represents the nominal cleavage strength of a steel: the latter is defined as the highest nominal stress a steel can sustain in the presence and under the influence of a notch caused by a crack and for particular conditions of testing. In effect, the NC curve expresses the nominal stress necessary for the initiation of a brittle fracture as a function of temperature and for the other prescribed conditions.

It is important to recognize that a fracture very seldom starts in an NC test-bar immediately after a crack occurs in one of the welded edges of the bar. Therefore, the observed nominal stress is not the stress that propagates the crack through the tested material, but represents the nominal stress for initiating a

fracture at the front of a sharp notch that is almost undeformed.

Another important assumption on which the method of testing is based is that the energy developed, when the weld cracks, is so small that it gives no essential addition to the applied external nominal tensile stress.

From an examination of the brittle fracture of steels and the characteristics of a steel, to arrest a running crack, it is evident that (a) a certain minimum stress must be applied to a steel for an existing fracture to propagate any considerable distance; (b) another minimum stress, usually greater than (a), must be applied to initiate a brittle fracture.

It is now clear that the minimum nominal stress for either purpose need not arise purely from external loading but may be completely or partly due to elastic prestresses, e.g. internal welding stresses, other residual stresses, or high elastic stresses caused by an elastic shock wave developed by a propagating brittle fracture. The maximum value of the last-mentioned type of stress occurs just in front of the running crack.

Elastic prestresses of residual type may be regarded as either a pure stress addition to an external service stress or a cause of an increased degree of triaxiality or complicated effects of both. This leads to the observation that a brittle fracture may be initiated as well as propagated over a considerable distance even if no external load has been applied to the material. In this case the internal stress must reach and maintain the required level without any external addition.

It is far from rare that brittle fractures occur in unloaded welded structures merely as a consequence of the welding stress level, if a sharp notch, e.g. a defect in a weld, is present. It is important to record that cracks formed in this way usually have a very short "deadening distance". This may be interpreted as indicating that the elastic shock wave formed at the fracture front cannot maintain the required stress level for more than a short time after the internal stress has been relieved. The crack will therefore stop when it reaches parts of the material in which normal internal and external stresses are absent.

For a particular nominal stress, it has al-

ready been stated that the initiation of a brittle fracture at the root of a sharp notch can only be presented by plastic deformation at the front. This critical stress condition decreases with decreasing temperature. However, there is another factor which strongly influences the possibility of plastic deformation at a notch-front, namely the strain rate<sup>21</sup> - the higher the strain rate, the less the plastic deformation. The translation, the only type of deformation which is of practical importance, is strongly time-dependent. For example, the rapidly increasing yield strength of a steel with increasing high strain rate in conventional tensile testing, is well known. Thus at sufficiently high strain rate, a brittle fracture can also be initiated by a nominal stress which falls well below the NC-test nominal cleavage strength at a particular temperature - the greater the shock effect of high strain rate, the lower the nominal stress level to cause a cleavage fracture. The latter statement, however, seems only to be valid down to a certain limit characterizing each particular steel.

There are certainly experimental difficulties in the accurate determination of this limit for a steel, but a reasonable approximation appears to be provided by a feature of the NC curve. It is known that in the temperature range of -150 C to -200 C almost completely brittle cleavage fractures occur even with uniaxial stress application. Slight plastic deformation may be unavoidable in tensile tests made within this temperature interval or at still lower temperatures, but from a practical point of view this unavoidable deformation can be regarded as negligible. The above temperature range may therefore be associated with nil plastic deformation of a steel. Consequently, no plastic deformation can occur at a crack front within this range. Thus, if an extrapolation of the straight line given by an NC curve on logarithmic scale is made down to say -200 C, a value of the nominal cleavage strength is obtained which is the lowest nominal stress that can cause a cleavage fracture at the front of a crack. This limit value, here called  $\sigma_p$  (p for propagation) is nothing but the nominal cleavage strength when for all practical purpose any form of deformation can be excluded.

-----  
It is quite evident that a typical condition

of stress in a steel when negligible plasticity occurs at a crack front, is associated with a propagating brittle fracture. The essential factor that prevents plastic deformation is the high strain rate, since the influence of the strain rate predominates that of temperature over a wide temperature range. Cleavage fractures can progress as soon as the nominal elastic stress exceeds a certain level. A well-founded opinion is that the nominal stress level required is approximately the same as the graphically determined value.

It is important to note that initiation and propagation of a brittle fracture apply to the same phenomenon, and the term "propagation" can be replaced by "continuous initiation". There is experimental evidence, which confirms that propagation can be interpreted in this way.

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The  $\sigma_{c,}$ -level may be regarded as an approximate value of the nominal stress which is the minimum demanded for continuous initiation of a brittle fracture (propagation).

-----

The NC-testing method has been described in several other papers<sup>2-22-23-24</sup> and the foregoing description might be completed by quoting another one:<sup>2</sup>

The above-mentioned will underline the importance of the slope of the NC curve - the steeper the curve the lower its critical stress for propagation of a brittle fracture.<sup>2-6</sup> According to the NC-testing method stress levels necessary for brittle fracture propagation as low as about 0.5 kg/mm<sup>2</sup> have fairly frequently been found for ordinary carbon steels. However, even if such low values may be regarded as exceptions, unalloyed structural steels with critical nominal stress levels for brittle fracture propagation exceeding 7-8 kg/mm<sup>2</sup> are not often observed.

Further aspects on the slope of the NC-curve and its importance have been discussed in previous papers.<sup>7-8</sup> To summarize, a steel for welded structures should have a low critical temperature  $T_c$  and an NC curve with the smallest possible steepness.

The importance of the temperature  $T_c$  will

appear more clearly when it has now been experimentally proved that below this temperature welding stresses may have a most detrimental influence on the stability of a welded structure, while they at temperatures above  $T_c$  do not seem to have the same serious effect, if any. A detailed discussion of the effect of welding stresses<sup>3-7-8</sup> should not be repeated here but only exemplified by Fig. 34. It shows that under influence of welding stresses the part of the NC curve below  $T_c$  has no practical importance. Consequently, below this temperature only the stress level necessary for propagation of a brittle fracture has to be considered, while an initiation of such a fracture may be caused by any service load added to residual welding stresses, if the total stress level will then reach the actual NC curve at a certain temperature.

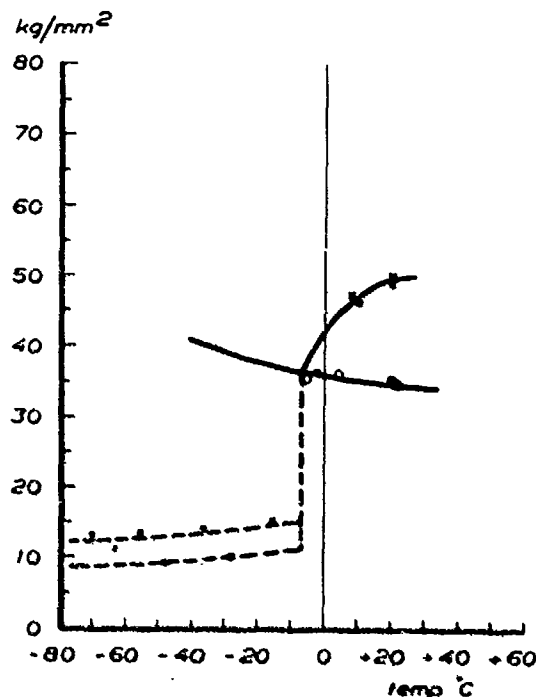


FIG. 34. EFFECT OF WELDING STRESSES ON THE NC DIAGRAM. THE PART OF THE NC CURVE BELOW THE TEMPERATURE  $T_c$  (THE INTERSECTION POINT BETWEEN THE NC CURVE AND THE YIELD STRENGTH CURVE) WILL NO LONGER EXIST AND THIS PART OF THE CURVE WILL BE REPLACED BY A CURVE WHICH SHOWS THE CRITICAL STRESS LEVEL FOR PROPAGATION OF A BRITTLE FRACTURE. APP. B.

This can also be expressed by saying that

residual stresses alone, caused for example by welding, can reach a level high enough for initiation of a brittle fracture, while such a fracture will never propagate, if the nominal stress level due to service load is not sufficient. Hence, if a brittle fracture has become initiated due to residual welding stresses it will run through the structure only so far a distance that will be permitted by the energy release and the area under influence of welding stresses, the so-called deadening distance.<sup>7-8</sup>

In order to complete this way of considering brittle fractures in steel with respect to stress levels, the interpretation may be extended to include the possibilities for a steel to act as a crack arrester also below the transition temperature  $T_c$ . Without going into details this can be summarized by saying that the stress level caused by service loads in relation to the nominal cleavage strength of the steel is the factor, which is determining the crack arresting properties of a steel below  $T_c$ . If the service stress is close to the nominal cleavage strength at the temperature in question, the crack arresting effect is small and the deadening distance of the crack considerable. If, on the other hand, the nominal stress acting on the structure is lower or only a little above the  $\sigma_{c,p}$ -level the crack may be arrested also below  $T_c$ .

This statement can be applied also to the behavior of a fracture when it is running through a structure, passing welds and plates. In and around welded joints the stress level is always relatively high, while the nominal service stress in parts, which are not influenced by welding stresses, may be rather low. Consequently the energy absorption will also be very low, when the crack passes through a welded joint where the chevron pattern of the fracture surface is not very distinct. A higher degree of energy absorption will be found when the crack runs through parts of the structure, which are subject to a low-stress level and the fracture surface shows a pronounced chevron pattern. The pattern will become still more pronounced when the propagation rate of the crack is decreasing. The conditions for propagation of a brittle fracture through a structure must therefore be dependent upon a combined effect of

1. the energy contents of the propagating fracture,

2. the nominal cleavage strength of the steel,
3. the  $\sigma_{c,p}$ -level of the steel,
4. the total stress level in the structure, combined by the nominal service stress and residual stresses.

-----

The function stability of a welded structure is determined by two factors,

1. the conditions necessary for initiation of a fracture,
2. the conditions necessary for propagation of a brittle fracture.

The conditions to be fulfilled for initiation of a fracture is a total nominal stress exceeding the nominal cleavage strength, and a temperature, which is below the critical temperature  $T_c$ . For propagation of a brittle fracture the corresponding conditions are a nominal stress higher than the one, which is critical for brittle fracture propagation and a temperature below the upper transition temperature  $T_s$ .

-----

It must be underlined once more that an initiation of a fracture according to the nominal cleavage strength diagram necessarily needs a very sharp notch, i.e. normally a natural crack. What is generally called a sharp notch, e.g. a Charpy V-notch is not sufficiently sharp to illustrate the nominal cleavage strength of a steel.

-----

With the background given it is now possible to describe how welded joints will offer various degrees of function stability to a welded structure. The expressions unconditional and conditional instability, metastability, quasi stability and stability will be used.

Thus

- a. provided that initiation as well as propagation of a brittle fracture can take place, the structure is unconditionally unstable,
- b. provided that propagation but not initiation of a brittle fracture can take place the struc-

ture is metastable, i.e. in practice function stable on static ("resting") notch effects (and of course in the absence of such effects) but conditionally unstable on dynamic notch effects (i.e. in the latter case when attacked by a running brittle fracture from surrounding parts of the structure),

c. provided depending conditions for propagation but not for initiation exist the structure is quasi stable (see below) and

d. provided conditions neither for initiation, nor for propagation exist the structure is stable.

The above-mentioned statements a-d need some comments. In the presence of a sharp notch such as a crack or a similar defect in a part of the structure the circumstances shown by Fig. 35-36 are in principle valid. This can be exemplified by the following:

An unconditionally unstable part of a structure may be where hot cracks, hardening cracks, flakes, shrinkage cracks or the like have formed, the nominal service stress of which at certain service temperatures exceeds the nominal cleavage strength  $\sigma_c$ , in material free from residual stresses (= the area L in Fig. 35) and impact loads cannot be excluded.

On sufficiently high residual stresses, for instance welding stresses, the part of the welded structure is unconditionally unstable already if it is subject to temperatures below  $T_c$  and a service load is not necessary for initiation or propagation of a brittle fracture (= the area L in Fig. 36).

A metastable part of a structure is illustrated by M in Fig. 35-36. In this case the nominal stress is lower than the  $\sigma_c$ -curve and the presence of "resting" crack notches ("static-notch effect") will not lead to any brittle fracture risk. The area M of both diagrams exceeds, however, the  $\sigma_{cp}$ -curve (= the nominal stress necessary for brittle fracture propagation, "continuous initiation"). On "dynamic-notch effect", i.e. if the part of the structure will become attacked by a running brittle fracture, stresses within the area M are sufficiently high for the fracture to proceed. The metastability will then change into instability, but this is conditional. Since there

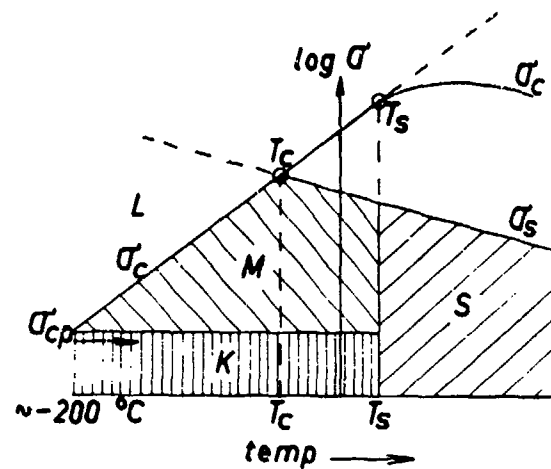


FIG. 35. COMPLETE NC DIAGRAM, WHICH SHOWS IN PRINCIPLE THE YIELD STRENGTH. THE NOMINAL CLEAVAGE STRENGTH (BOTH AS STRAIGHT LINES IN THE LOGARITHMIC STRESS SCALE) AND THE AREAS WITH VARIOUS STABILITY ACCORDING TO THE TEXT. THIS DIAGRAM REPRESENTS THE STRESS RELIEVED CONDITION OF A STEEL. FURTHER SYMBOLS REPRESENT YIELD STRENGTH ( $\sigma_s$ ), NOMINAL CLEAVAGE STRENGTH ( $\sigma_c$ ), CRITICAL PROPAGATION STRESS FOR BRITTLE FRACTURES ( $\sigma_{cp}$ ), LOWER TRANSITION TEMPERATURE ( $T_c$ ) AND UPPER TRANSITION TEMPERATURE ( $T_s$ ). APP. B.

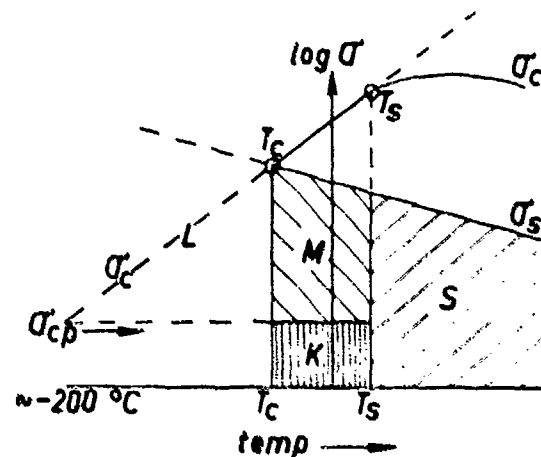


FIG. 36. SAME TYPE OF DIAGRAM AS IN FIG. 35 BUT REPRESENTING A STEEL UNDER THE INFLUENCE OF RESIDUAL WELDING STRESSES. THE AREAS OF VARIOUS STABILITY ARE MARKED ACCORDING TO THE TEXT. FURTHER SYMBOLS REPRESENT YIELD STRENGTH ( $\sigma_s$ ), NOMINAL CLEAVAGE STRENGTH ( $\sigma_c$ ), CRITICAL PROPAGATION STRESS FOR BRITTLE FRACTURE ( $\sigma_{cp}$ ), LOWER TRANSITION TEMPERATURE ( $T_c$ ) AND UPPER TRANSITION TEMPERATURE ( $T_s$ ). APP. B.

is always a certain arrest effect on a brittle fracture on nominal stress levels lower than  $\sigma_c$  (caused by a small but still plastic deformation adjacent to the fracture surfaces) the instability will depend on the energy contents of the fracture and the stress level within the area M. The higher these are the more the instability will approach the unconditional one (the  $\sigma_c$ -level). Even if a certain crack-arresting effect caused by energy absorption within the area M will take place in connection with a brittle fracture, one must always use the terms metastability and instability for such stress levels, since the deadening distance is always considerable, if a brittle fracture has once become initiated.

The term "quasi stability" is used as an expression for a stability that is not quite true and complete.

A quasi stable part of a structure (stress and temperature area K in Fig. 35-36) is such a part, which theoretically should be stable because of the fact that the nominal stress is below the  $\sigma_s$ -level, but which is subject to such a low temperature that considerable failures may still occur through a sudden brittle-fracture attack. One can imagine a welded joint in which, owing to welding stresses and a notch effect, a brittle fracture has become initiated. Let us further assume a low nominal stress in surrounding parts (below  $\sigma_s$ ). Under these conditions the surrounding parts are quasi stable. The crack will not propagate further than the deadening distance, but this distance depends upon the energy release in connection with the initiation. This can seldom be foreseen and the quasi stable state is nothing to rely upon. A practical example is a welded structure with a heavy plate thickness under fabrication. Should a brittle fracture become initiated in an almost finished weld caused by welding stresses and defects, the deadening distance will be long because of the high-energy release and the whole structure may be spoiled in spite of no additional service load. However, should a corresponding fracture occur in the structure, but at the beginning of the welding, the deadening distance may be only an inch or two. Consequently the brittle fracture will become arrested, since there is no service load acting as a driving force.

In both cases the parts around the welded joint are quasi stable, i.e. theoretically

stable, but the conditions in connection with the initiation of the fracture which may cause this stability under practical circumstances is not quite true and complete.

One may now return to a publication by Kochendörffer and Scholl.<sup>28</sup> Also the NC-

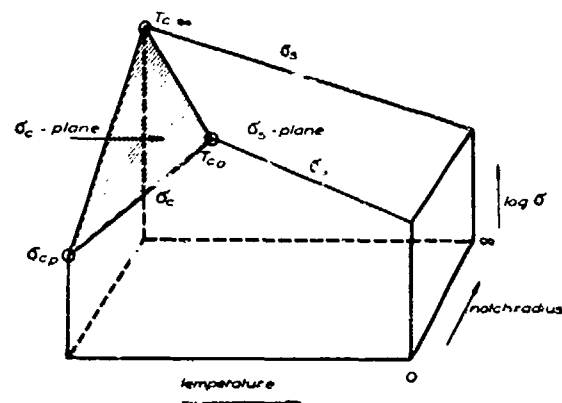


FIG. 37. THREE-DIMENSIONAL NC DIAGRAM, WHICH SHOWS IN PRINCIPLE THE CONDITIONS FOR BRITTLE FRACTURE INITIATION VS. STRESS (LOGARITHMIC SCALE), TEMPERATURE AND NOTCH RADIUS. FURTHER SYMBOLS REPRESENT YIELD STRENGTH ( $\sigma_s$ ) AND NOMINAL CLEAVAGE STRENGTH ( $\sigma_c$ ). APP. B.

testing results can be illustrated by the three-dimensional diagrams schematically shown in Fig. 37. This diagram with a logarithmic scale on the vertical stress axis shows the yield strength and the cleavage strength planes and how they are depending of the third variable, the notch effect.

### APPENDIX C

#### SUMMARY OF RECENT INVESTIGATIONS

The result of additional investigations\* on the properties of niobium-(columbium) treated mild steel can be summarized in the following:

1. The solubility of niobium in austenite could approximately be expressed by the equation

$$\log (\% \text{ Nb}) (\% \text{ C}) = -2500/T - 0.63$$

For example with 0.20% C the solubility at a normal reheating temperature (2200°F) is slightly above 0.02% Nb.

2. The increased strength could be mainly attributed to the precipitation of niobium, probably as carbide, with its maximum effect at 1200°F. Coarse precipitates, which remained undissolved during soaking, did not contribute to any significant extent to mechanical properties.

3. The transition temperature was raised mainly because of precipitation hardening.

4. Annealing at temperatures above 1200°F caused softening because of coarsening of the precipitate. The effect of precipitation hardening on mechanical and impact properties disappeared after annealing at about 1450°F.

5. Hardness increase after quenching from solution treatment temperatures and subsequent annealing at 1200°F was 4-5 times higher than after continuous cooling. This indicates that during continuous cooling some precipitation occurs at higher temperatures. No significant precipitation, however, was observed in the austenite when all niobium was brought into solution. When part of the precipitate remained undissolved, some dissolved niobium was precipitated also in the austenite.

6. Cooling rate in the range 20° to 150°F/min, measured between 1290° and 1110°F, had

very small effect only on mechanical and impact properties. Within these limits it did not seem to affect the precipitation itself.

7. Precipitated niobium inhibited austenite grain growth up to higher temperatures than is normally observed in aluminum-killed steels. Normalized niobium-treated steel retained its mechanical and impact properties after overheating up to about 1900°F.

8. The ferrite grain size of normalized niobium-treated steel was on an average ASTM No. 10.3, which is smaller than of normal aluminum-treated steels. The fine grain size resulted in increased yield stress and improved notch ductility, with a Charpy-V 20 ft-lb transition temperature around -80°F.

9. According to an investigation by Ronn\*\*, dissolved niobium retarded the transformation into pre-eutectoid ferrite and pearlite--the former resulting in a marked tendency to Widmanstätten structure formation in continuously cooled steel. Notch ductility is impaired if the structure contains substantial amounts of Widmanstätten ferrite.

10. In the same investigation niobium was not found to have any substantial effect on the transformation into bainite. The bainite formation occurs as in the corresponding base alloy without any niobium addition.

11. Rolling temperature between 1470° and 1830°F did not affect mechanical and impact properties when the reheating temperature before rolling was 2370°F. A beneficial effect of low rolling temperatures (controlled rolling) on impact properties, however, was observed with a reheating temperature of 2200°F. In the former case the entire structure consisted of

Widmanstätten ferrite and bainite or pearlite at all rolling temperatures. In the latter case the major part of the ferrite was equiaxed, and the occurrence of Widmanstätten ferrite decreased with decreasing rolling temperature. It is therefore believed that the beneficial effect of controlled rolling can mainly be attributed to the formation of smaller austenite grains, and that the presence of some undissolved precipitate is a condition for this to occur.

\* de Kazinczy, F., Axnas, A., and Pachleitner, P., "Some properties of niobium-treated mild steel," *Jernkontorets Annaler*, 147:4 (1963), p. 406.

\*\* L. Ronn, Graduate Thesis Work at the Royal Institute of Technology, Stockholm, 1962.



12. The addition of small amounts of Cr and Mo to niobium-treated steel increases the austenite grain size, whereas the addition of Al or Ti decreases it. This was reflected among other things in decreased yield/tensile ratio in the former case, and in increased ratio in the latter one, both in the as-rolled and normalized condition. Notch ductility was affected in the same manner.

13. The critical stress for brittle-fracture initiation according to the Orowan concept was decreased by niobium in the as-rolled condition, but was restored after strain-aging or normalizing. Niobium-treated steels, finished in the higher temperature range, did not exhibit a Lüders strain and revealed virtually no difference between upper and lower yield points. This phenomenon also disappeared after strain-aging or normalizing. Increased density of mobile dislocations in the as-rolled condition is offered as a common explanation.

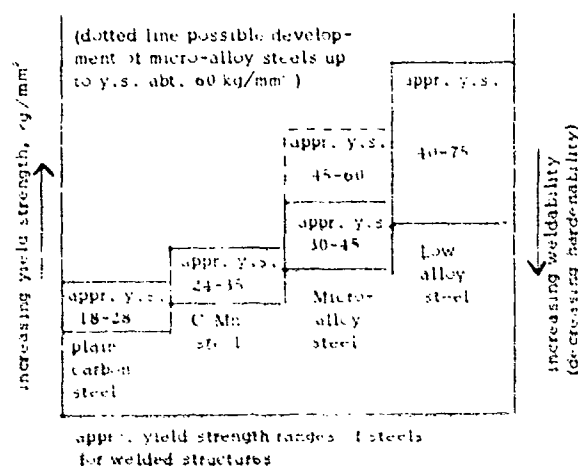
14. Niobium delays strain-aging in the temperature range of nitrogen aging by a time factor of 4 both in the as-rolled and normalized condition. The maximum increase in yield stress was also somewhat reduced. A possible explanation is that some nitrogen is precipitated with niobium, thus reducing the content of dissolved nitrogen.

In an additional investigation by Noren and de Kazinczy<sup>3</sup>, previously mentioned in the paper, the embrittlement of various steels upon stress relieving in the temperature range between 930° and 1290°F was studied. It was found that the Charpy-V transition temperature of a semikilled normalized niobium-treated steel increased less than that of aluminum-treated steels - the difference being particularly pronounced after annealing for 24 hr at 1200° and 1290°F.

Further additional investigations have been performed concerning the weldability of niobium-treated steels. These investigations include a great number of various experiments, in laboratory scale as well as with the steel in full-scale welding production. They may be summarized by the statement that the weldability of niobium-treated steels within the composition ranges mentioned in the paper has proved to be superior to carbon-manganese steels and even plain carbon steels in plate thicknesses equivalent in strength.

Finally, promising results have been obtained on investigating more complex micro-alloy steels of increased strength in which niobium is one of the micro-alloying elements. Still more promising experiences as to the possibility of reaching considerably higher strength levels of micro-alloy steels after proper heat treatment have appeared, but for the present it is too early to report any details.

Our present feeling concerning the application of micro-alloy steels for welded structures, and particularly with regard to those in which niobium alone is the micro-alloy addition, is rather optimistic as to the strength levels discussed in the paper as well as to micro-alloy steels with still higher strength. This can be illustrated by a diagram showing, in principle, the relation between yield strength and weldability for various steel groups:



The diagram shows approximately the yield strength ranges covered by plain carbon steels, C-Mn-steels, various micro-alloy steels and low-alloy steels for welded structures. On increasing strength within each group of steels, the weldability will become impaired, mainly because of higher carbon contents and/or contents of other alloying elements, i.e. due to increased hardenability. This will imply, for comparable strength levels of two steel groups, that it is favorable from weldability point of view to choose the steel, which will fulfill the strength specifications within the lower part of the possible strength range of the group rather than a steel that has to be chosen within the upper part of its range for the same strength. Hence there is obvious a good

deal to be gained by using a micro-alloy steel rather than a C-Mn-steel as, for example, high-strength ship steel, provided the latter has to be produced with carbon and manganese contents very close to the acceptable maximum figures to be permitted for welding. A micro-alloy steel with corresponding yield strength but lower in carbon and manganese will certainly withstand much more rough treatment in the shipyard.

#### APPENDIX D

##### EXTRACT OF INVESTIGATION FOR THE OFFICIAL APPROVAL OF COLUMBIUM MICRO-ALLOY STEEL AS PRESSURE VESSEL MATERIAL ACCORDING TO REQUIREMENTS OF SWEDISH AUTHORITIES

The Swedish authorities have approved the use of columbium micro-alloy steel in the silicon-killed and normalized condition for use in pressure vessels and an extensive investigation has been performed for this approval. The Figures 38-48 show some results, which may complete the foregoing report with respect to impact strength, weldability and yield strength at elevated temperatures, representative for this type of steel. Various heats of columbium micro-alloy steel were investigated and the one represented by the Figures 38-48 had the following composition:

C 0.16 Si 0.32 Mn 1.42 P 0.026 S 0.017  
Cb 0.026

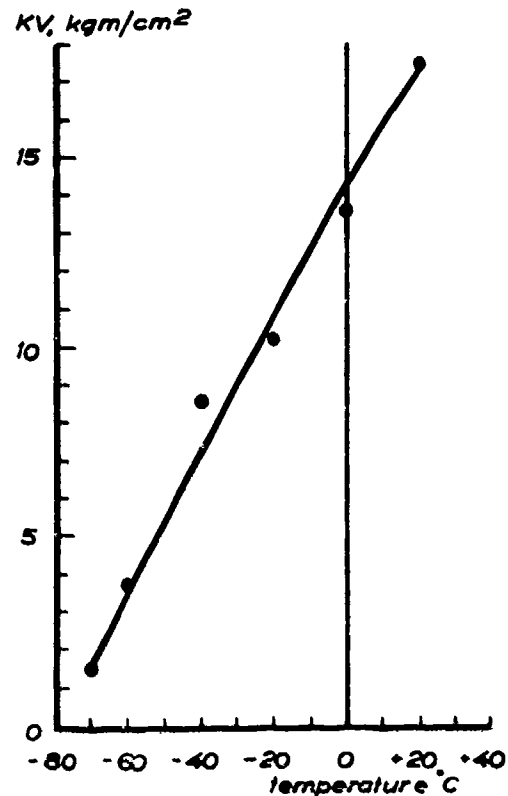


FIG. 38. CHARPY V-NOTCH IMPACT CURVE FOR 25 mm PLATE THICKNESS. APP. D.

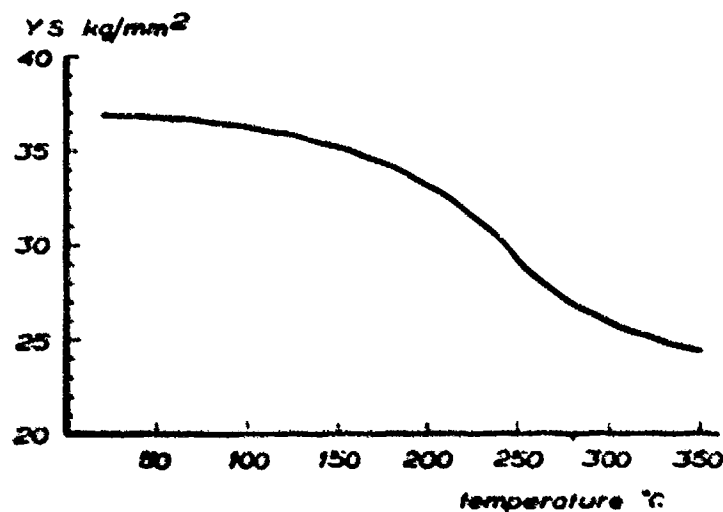


FIG. 39. YIELD STRENGTH VERSUS TEMPERATURE FOR A NORMALIZED SILICON-KILLED COLUMBIUM STEEL ACCORDING TO APP. D.

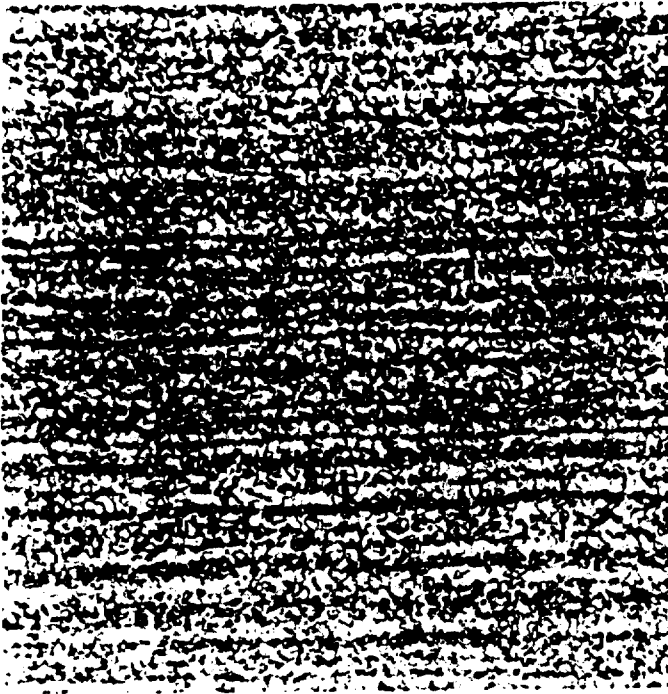


FIG. 40. MICRO-STRUCTURE IN 25 mm PLATE THICKNESS. APP. D. 100 x

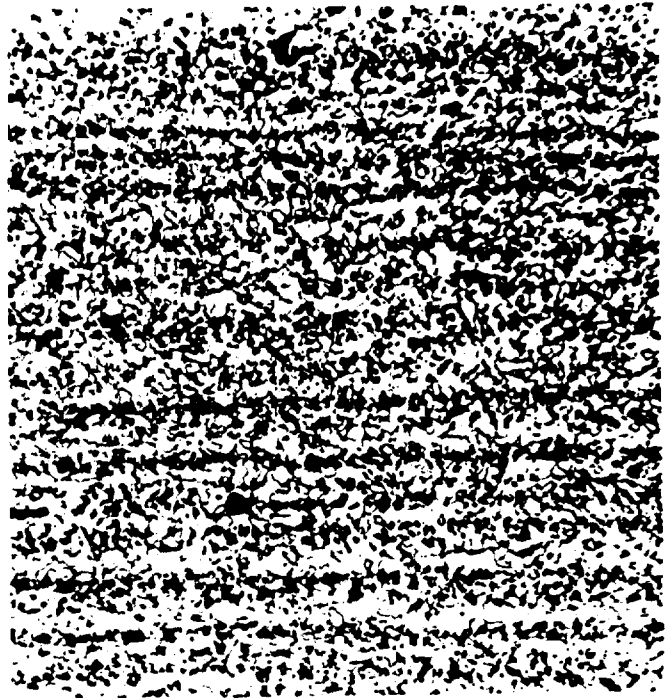


FIG. 42. MICRO-STRUCTURE IN 50 mm PLATE THICKNESS. APP. D. 100 x

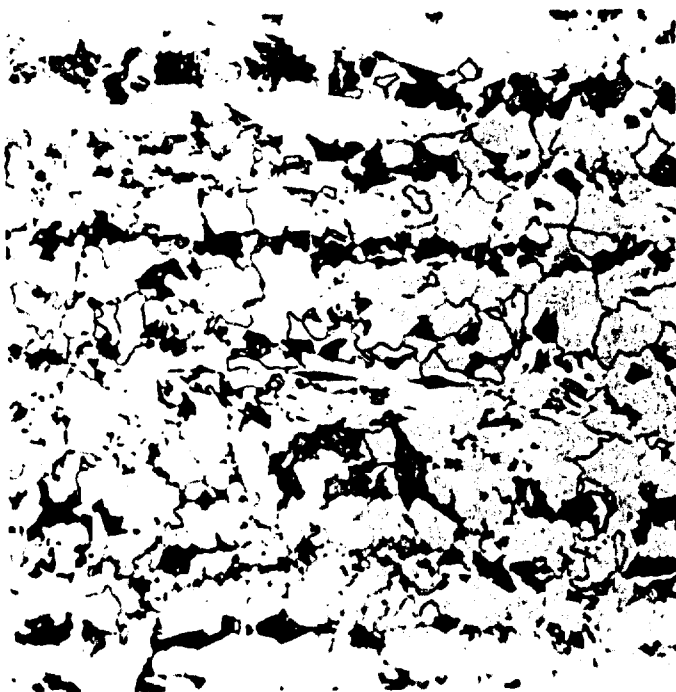


FIG. 41. MICRO-STRUCTURE IN 25 mm PLATE THICKNESS. APP. D. 400 x

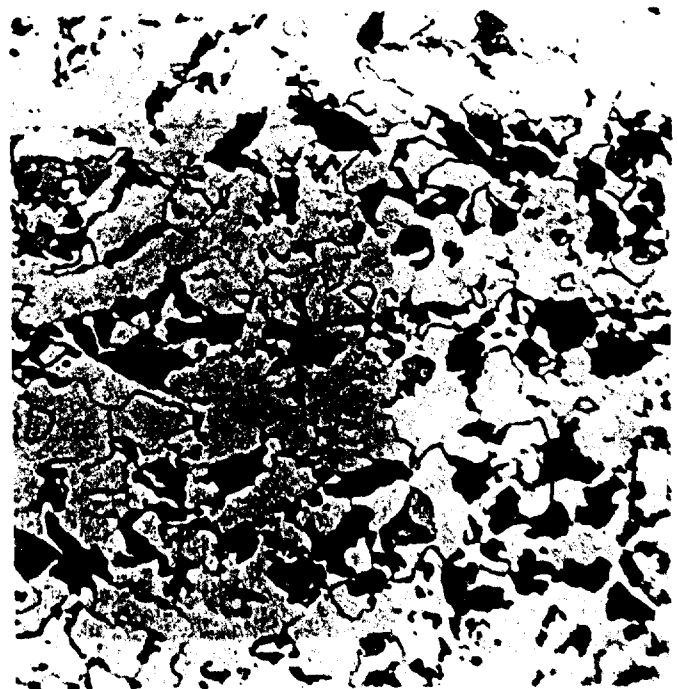


FIG. 43. MICRO-STRUCTURE IN 50 mm PLATE THICKNESS. APP. D. 400 x



FIG. 44. HEAT AFFECTED ZONE CLOSE TO FUSION LINE OF A WELD (4 mm ELECTRODE DIAMETER) AS A ONE-LAYER BEAD ON 50 mm PLATE THICKNESS. NO PREHEATING. MAXIMUM HARDNESS ABT. 325 HV. APP. D. 500 x



FIG. 46. OBSERVATION IN THE ELECTRON MICROSCOPE REVEALS THAT THE DENSER PARTS OF THE MICRO-STRUCTURE IN FIGURES 44 AND 45, WHICH ON NORMAL LIGHT MICROSCOPE LOOK AS MARTENSITE, ACTUALLY IS A TYPE OF LOW-TEMPERATURE BAINITE. APP. D. ELECTRON MICROGRAPH 12,000 x.



FIG. 45. HEAT AFFECTED ZONE CLOSE TO FUSION LINE OF A WELD (4 mm ELECTRODE DIAMETER) AS A ONE-LAYER BEAD ON 50 mm PLATE THICKNESS. NO PREHEATING. MAXIMUM HARDNESS ABT. 325 HV. APP. D. 1000 x



FIG. 47. IN THE ELECTRON MICROSCOPE THE COARSER PARTS OF THE MICRO-STRUCTURE IN THE FIG. 44 AND 45 ARE REVEALED AS A HIGH TEMPERATURE BAINITE. APP. D. ELECTRON MICROGRAPH 12,000 x

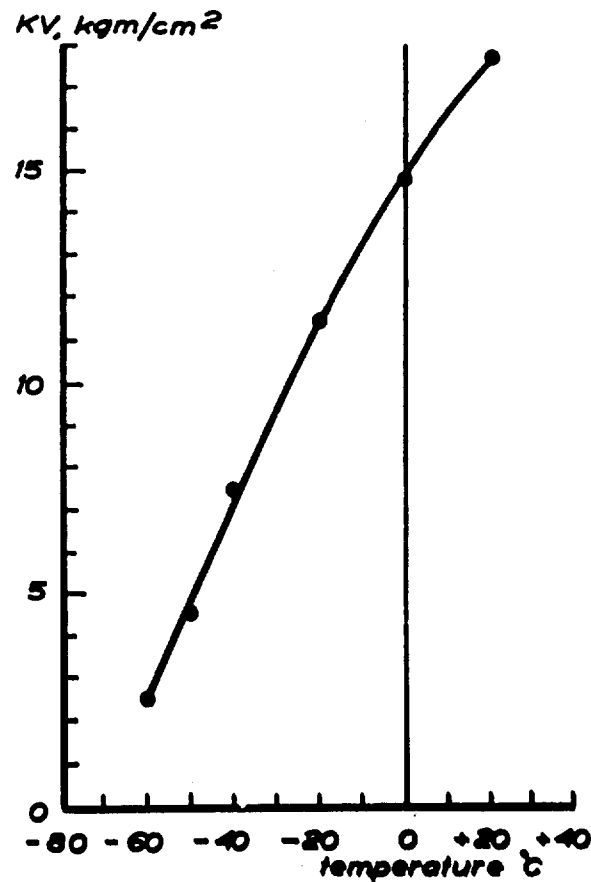


FIG. 48. CHARPY V-NOTCH IMPACT CURVE FOR 50 mm PLATE THICKNESS, APP. D.